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The Magazine of Metallurgical Engineering

INCLUDING

CURRENT METALLURGICAL ABSTRACTS

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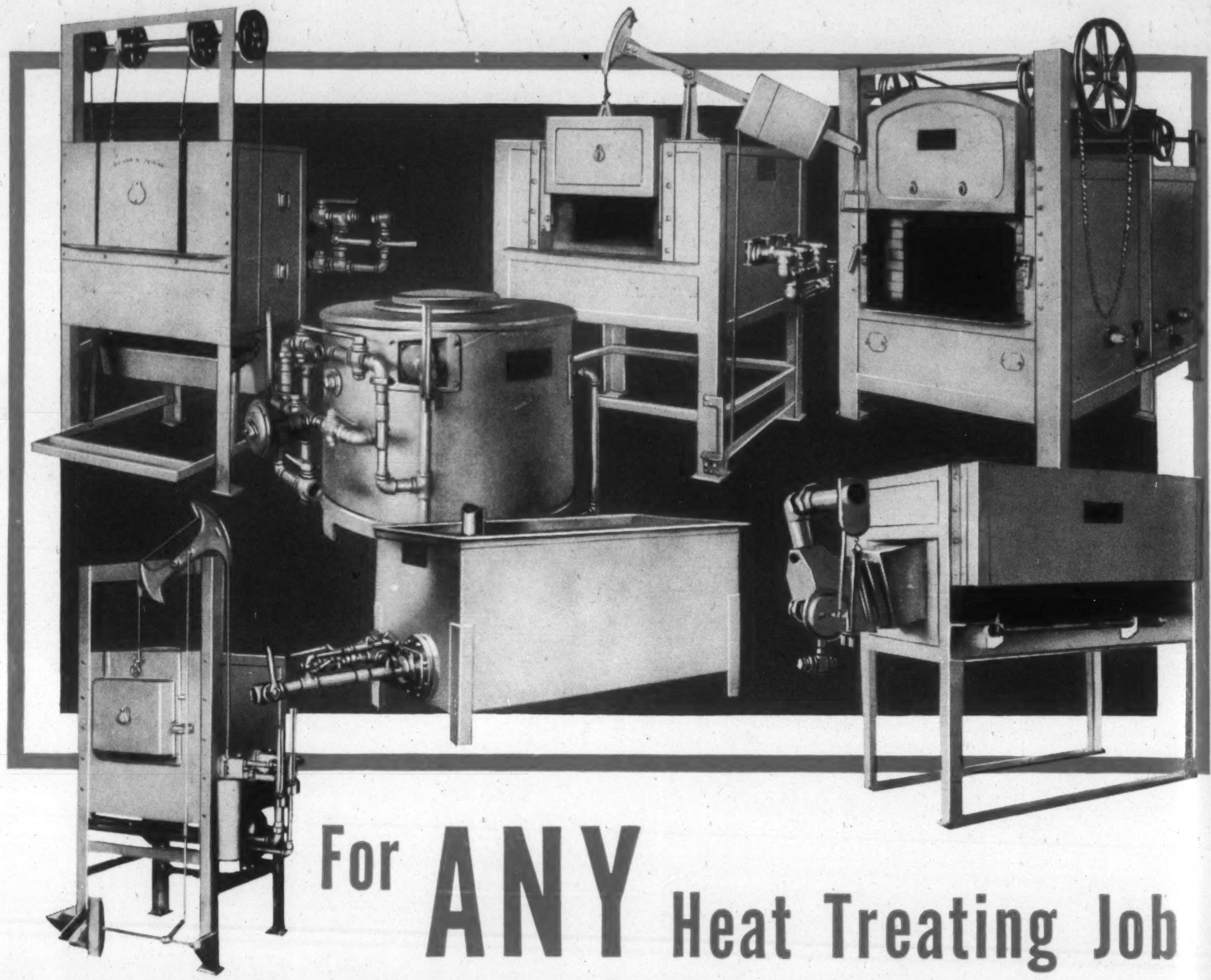

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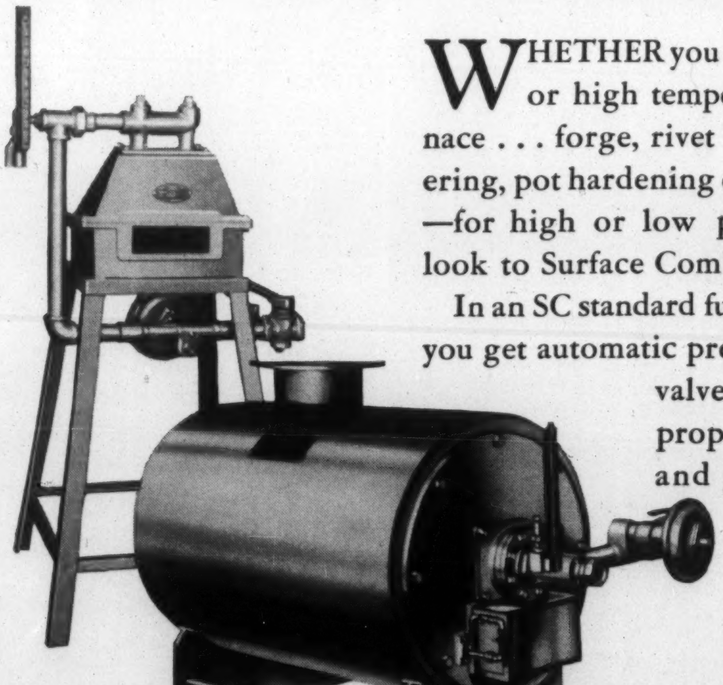

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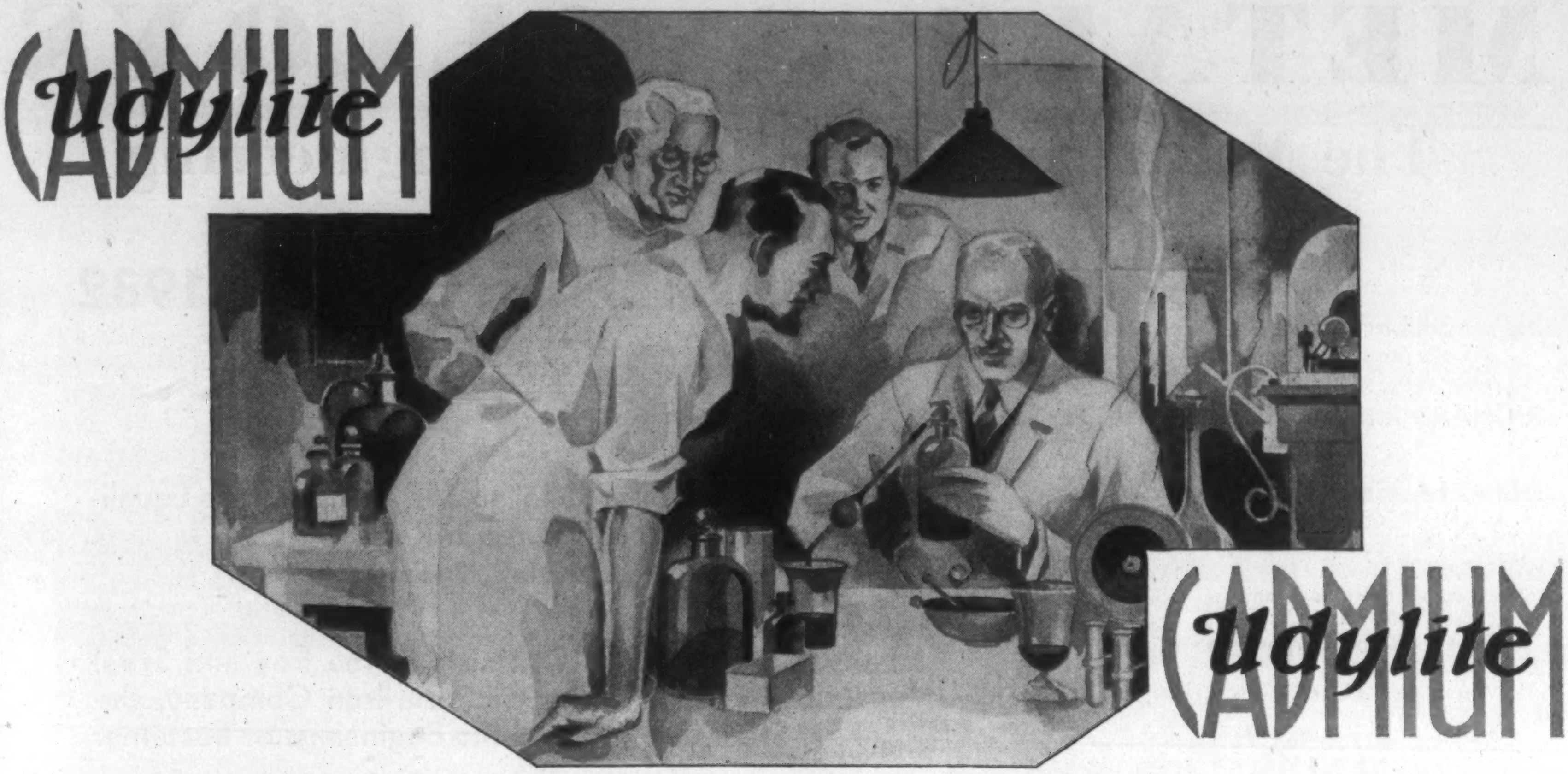

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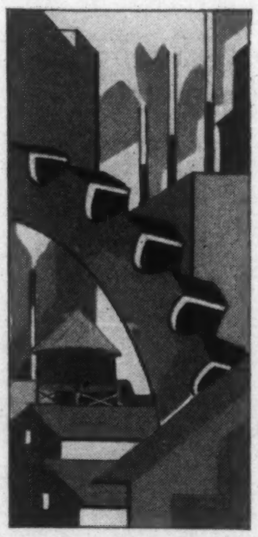

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EDITORIAL COMMENT

La Verne W. Spring

I HAVE been asked to do something which both honors and saddens me. I am honored in the request to pay a tribute for "Metals and Alloys," to LaVerne W. Spring, metallurgist of outstanding ability, and fellow-member with me on the Editorial Advisory Board of this publication.

I am deeply saddened, because of the loss to his profession, of a man of such signal, widely recognized ability; and because I was fortunate enough to be in Verne Spring's large circle of appreciative friends.

Many men formed a happy and helpful acquaintance with Spring, as members of technical society committees. There he was always an exponent of what was true as to fact, and what was fair to other people. He was among those who could always be depended upon for conscientiously fulfilling every obligation. His excellent judgment, supplementing his broad, unusually valuable experience, made him a tower of strength to every organization with which he was connected, and an individual who was eminently respected.

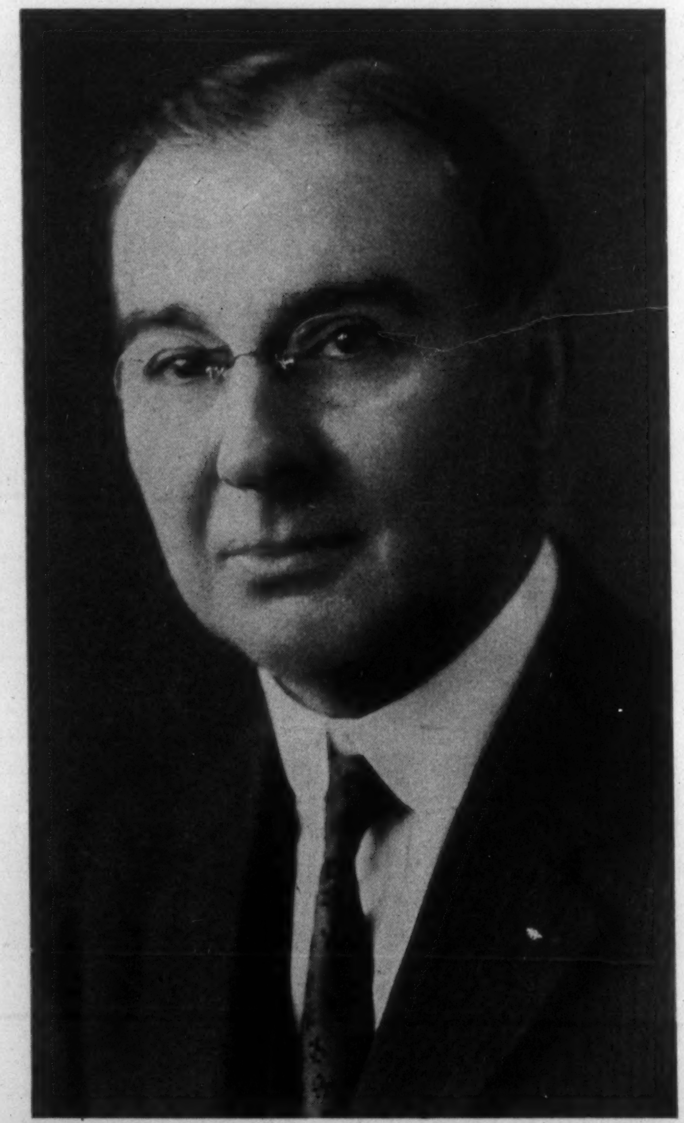
It seems to me that Verne Spring exemplified, to an extraordinary degree, an invaluable combination of industry, intelligence, tolerance, and honesty.

Spring's many friends now deeply sympathize with the members of his family, and with the company he served faithfully for many years. In these closer relationships as in many others, where his influence was periodically exerted, something fine survives.—R. A. Bull.

IN the passing of L. W. Spring, we have not only lost an outstanding metallurgist, we have lost an outstanding man. His technical achievements were of moment, but his point of view is of more moment. He gave freely of his time and energy to many metallurgical matters outside the confines of the interests of the Crane Company. Metallurgy was never a mere means of earning a living, in his eyes. He believed in scientific metallurgy, and to advance it was a gospel with him. Hence, he was active in metallurgical organizations and in committee work. Hence, his interest in Metals and Alloys and his service as a member of the Editorial Advisory Board. His "Non-technical Chats on Iron and Steel" and the series of articles on the Early History of American Metallurgy, upon which he and his colleague, Gilmore, were engaged at the time of his death, one of which appears in this issue, testify to the love he had for his profession.

In all our many contacts with him, we never once knew him to bow to expediency. When he took a position on any subject, it was the position dictated by strict adherence to truth. No apparent temporary advantage that might have been gained for himself or his firm ever influenced him in his vote or his discussion on a committee matter. Only the facts and his own convictions could affect his position. He stood four-square.

Many men have a wide circle of acquaintances, gained in their professional contacts. The unusually large number of contacts which Spring had professionally gave him not mere acquaintances but friends, for he was a friendly man, interested in the people he knew, not only as business and professional associates, but as people, so that all his associates naturally reciprocated. That wide circle of friends will not forget him. And the question, "How would Spring have looked at this matter or dealt with this problem?" raised at a committee meeting, will in the future tend to make those who are left take the same point of view he used to take. His influence will remain. H. W. Gillett



Business Insurance

ONCE upon a time a certain man became very sick. On his way to the hospital he canceled his insurance policy. Short-sighted? Poor business management? Foolish? Yes, but he was only following the example set by several large business concerns. One of the first steps in retrenchment made by some of our manufacturing companies was to reduce the research appropriation. Some companies abolished their research divisions. In a very real sense they canceled their insurance of business for the next period of good times.

An industrial research organization in a well managed company has two main functions; it must help the factory out of its current operating troubles, and it must develop new processes and new products by way of business insurance. Neither of these functions can be neglected if an industrial research laboratory is to do its duty by the company's stockholders. The job of "trouble

(Continued on Page 104)

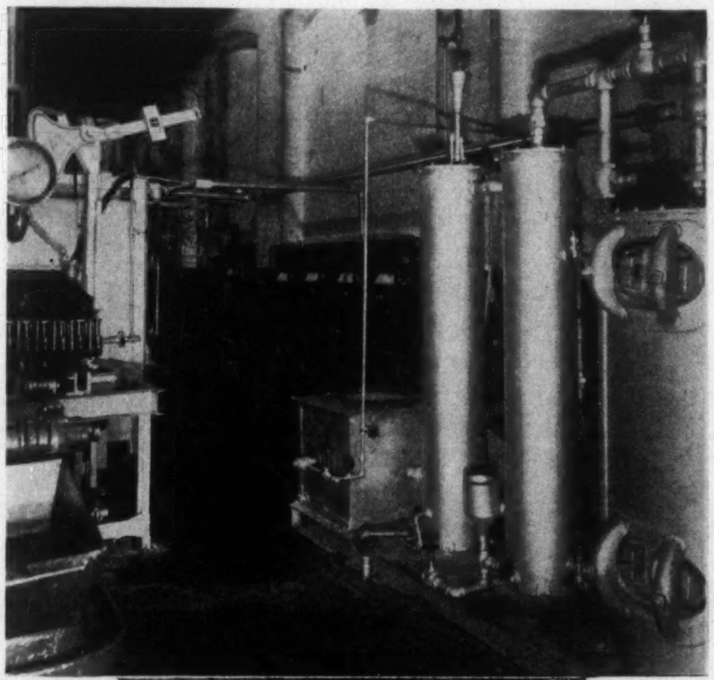
Continuous Gas Carburizing

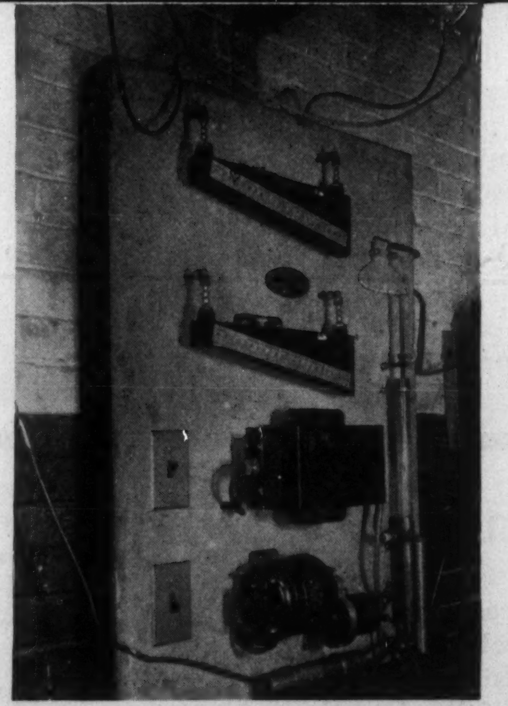
By E. A. THOMAS*

WHEN The Chrysler Corp. decided to undertake the manufacture of its own steering gears, it immediately became necessary to provide for the installation of the most suitable equipment for the heat treatment of the worms and worm wheels (sectors). From a metallurgical standpoint no serious difficulty was experienced in determining the proper types of hardening and draw furnaces to use for this work. Before installing a carburizing furnace, however, several vital factors had to be considered and might be listed as follows:

1. First cost.
2. Economy and simplicity of operation.
3. Cleanliness. This factor was very important in view of the fact that economy demanded the insertion of the entire heat treating unit in its proper place in the machine line.
4. Uniformity of case depth and carbon concentration. (Comparatively slight variations produce serious changes in size and spiral form.)
5. Rate of carbon penetration.
6. Facility for direct quenching from the carburizing furnace.

The management of the Newcastle Plant was impressed by the possibilities of a new type of continuous gas carburizer, designed by the Surface Combustion Corporation and, with the approval of the metallurgical department, contracted for the purchase of a furnace capable of handling 120 worms and sectors per hour. Installation was concluded the second week in December 1931 and three months of constant operation have vindicated the judgment of all concerned in its purchase.





Gas generating unit and its pressure and CO content control panel.

Theory of Carburizing

Little need be said here regarding the theory involved in this particular process of carburizing. It is essentially identical with that governing ordinary box carburizing. In fact, these fundamental chemical reactions were the basis on which this particular design was evolved.

The presence of CO at elevated temperatures, in intimate contact with the work is universally accepted as the primary requisite for all carburizing and is accomplished here by the reactions resulting from the introduction of a mixture of natural gas and moisture free flue gas into the muffle immediately beyond the charging valve. A uniform composition of this mixture is assured by the maintenance of a constant pressure ratio.

It appears, however, that the presence of a finely-divided carbon deposit on the surface of the steel is necessary prior to actual carburizing in order to produce the most satisfactory results. This deposition is coincident with the breaking down of a portion of the methane in the natural gas at a relatively low temperature in accordance with the following equation: $CH_4 = C + 2H_2$ and proceeds very rapidly due to the catalytic influence of the steel. During this reaction, which occurs in the first sections of the muffle, the CO_2 of the flue gas is inactive, performing the function of a diluent only.

As the temperature increases, the CO_2 of the flue gas reacts with the carbon deposited on the metal surface as follows: $C + CO_2 = 2CO$ and this brings about very rapid carburization of the steel.

These reactions continue throughout the entire cycle, although as the discharge end is approached and the relative proportion of CH_4 is reduced, the surface action decreases and carbon diffusion into the core is very pronounced.

Certain conditions must be constantly maintained to eliminate the possibility of the decarburizing action of H_2 and CO_2 at any stage in the process. Adequate dilution of the CO_2 by CO eliminates the deleterious action of the former, while the presence of even a small amount of CH_4 is sufficient to counteract the influence of the H_2. Both of these conditions are insured by the proper ratio of natural gas in the carburizing mixture.

The possibility of uncontrolled reactions in the final zone is eliminated by the continuous removal of used gases through a water-sealed muffle pipe.

The presence of any appreciable quantity of moisture not only reduces carburizing efficiency but has also a very harmful surface action on steel and is completely removed from the flue gas by means of a suitable drying unit.

Moisture-free flue gas is produced in the unit shown in the upper illustration on this page, which comprises:

1. A small flue gas generator in which natural gas is burned under neutral conditions, eliminating the possible presence of O_2.
2. A cooling tower filled with coke, through which the hot gas passes upward against a constant flow of cold water. Approximately 25% of the moisture contained in the gas is removed in this tower.
3. A second tower containing a circulating saturated solution of $CaCl_2$, where

---

*Metallurgist, Chrysler Corp., Newcastle, Ind.

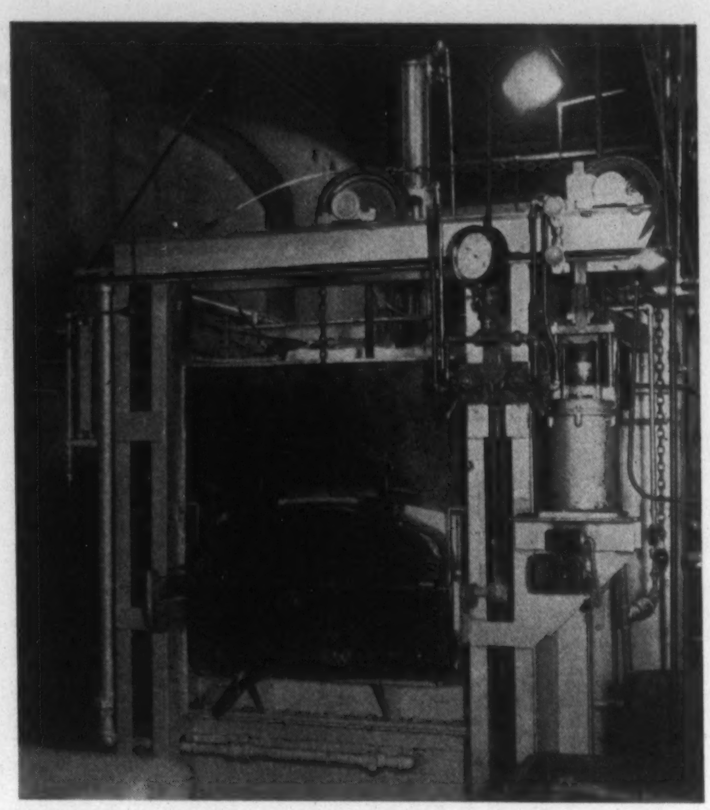
in Practice

approximately 90% of the remaining moisture is removed.

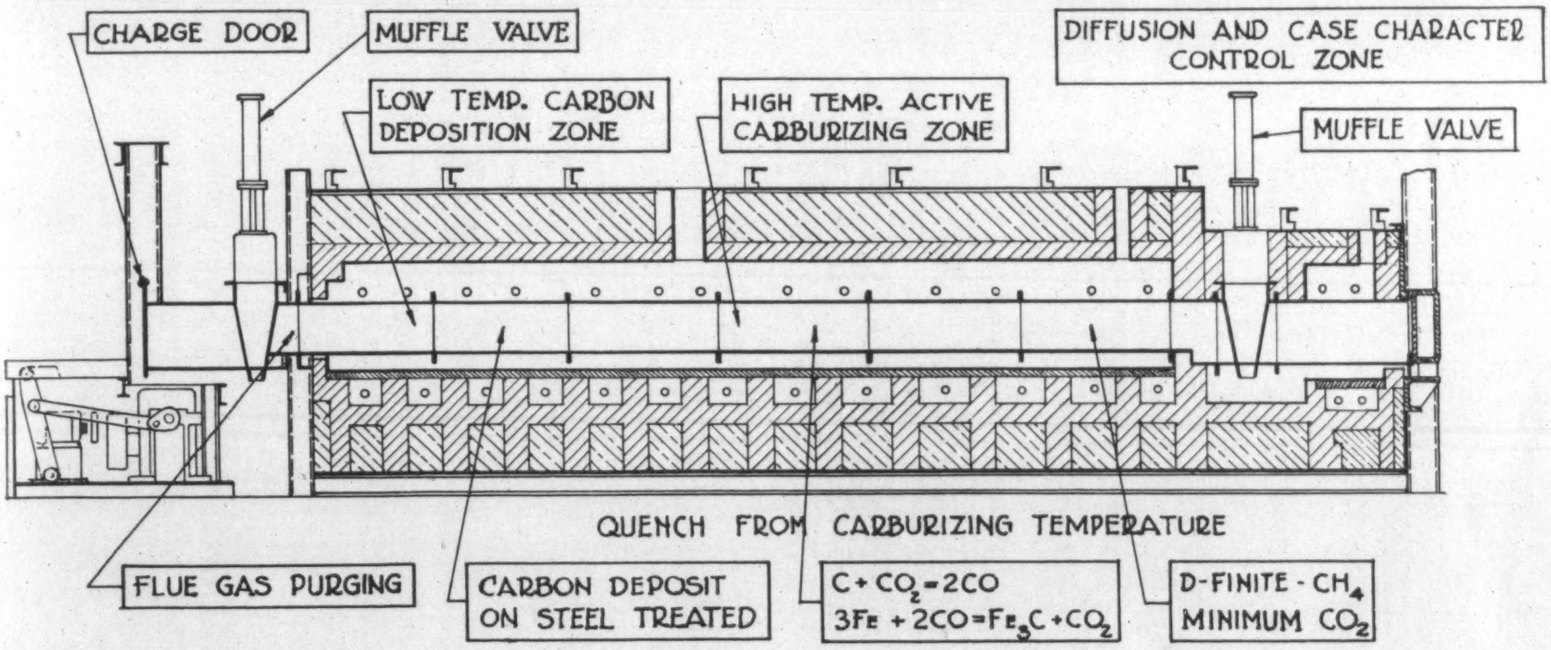
4. A third tower containing a mixture of dry $CaCl_2$ and wood shavings, which removes practically all remaining moisture.

Description of Furnace

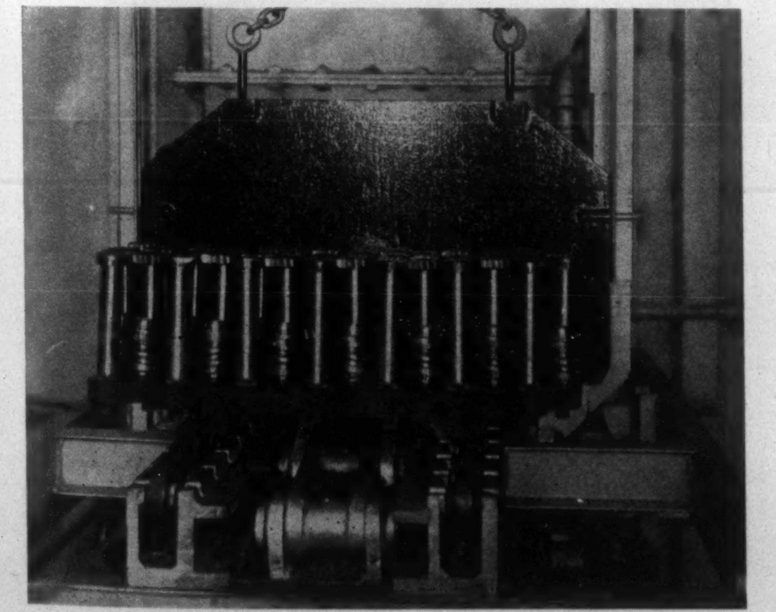
The furnace proper, the essential features of which are shown in the diagram herewith, is both over- and under-fired and contains an alloy muffle composed of flanged and bolted sections, which is anchored at the discharge end and floats towards the opposite end on alloy rollers supported in frequent alloy cross-supports. The carburizing trays are pushed through the muffle on two sets of rollers and skids by a suitable device. The trays and carburizing gas travel in the same direction. At each end of the muffle is an air-operated valve which acts as a seal. The valve at the charging end is air-cooled, while the discharge valve operates at approximately muffle temperature. The first two muffle sections, which comprise the deposition zone, are composed of a straight chromium iron, having no catalytic influence upon CH_4 decomposition. This eliminates carbon deposition on the walls of the muffle in this zone. All other sections are of a conventional Ni-Cr-Fe analysis. (In case at any time it should become necessary to replace a section, the muffle is disconnected at both valves and is raised and drawn out of the charging end of the furnace on rails.)



Discharge End and Control Valves.


CHARGE DOOR
MUFFLE VALVE
LOW TEMP. CARBON DEPOSITION ZONE
HIGH TEMP. ACTIVE CARBURIZING ZONE
DIFFUSION AND CASE CHARACTER CONTROL ZONE
MUFFLE VALVE
QUENCH FROM CARBURIZING TEMPERATURE
FLUE GAS PURGING
CARBON DEPOSIT ON STEEL TREATED
C + CO$_2$ = 2CO
3Fe + 2CO = Fe$_3$C + CO$_2$
D-FINITE - CH$_4$
MINIMUM CO$_2$

Schematic Longitudinal Section of Continuous Gas Carburizing Furnace.



Charging End. Steering Gear Worms and Sectors on Tray.

At the discharge end is located a holding chamber capable of maintaining any desired temperature in the work immediately prior to quenching. A purging chamber carrying two trays is located in front of the charging end. Flue gas is used for purging this chamber of air to eliminate the introduction of O_2 into the muffle.

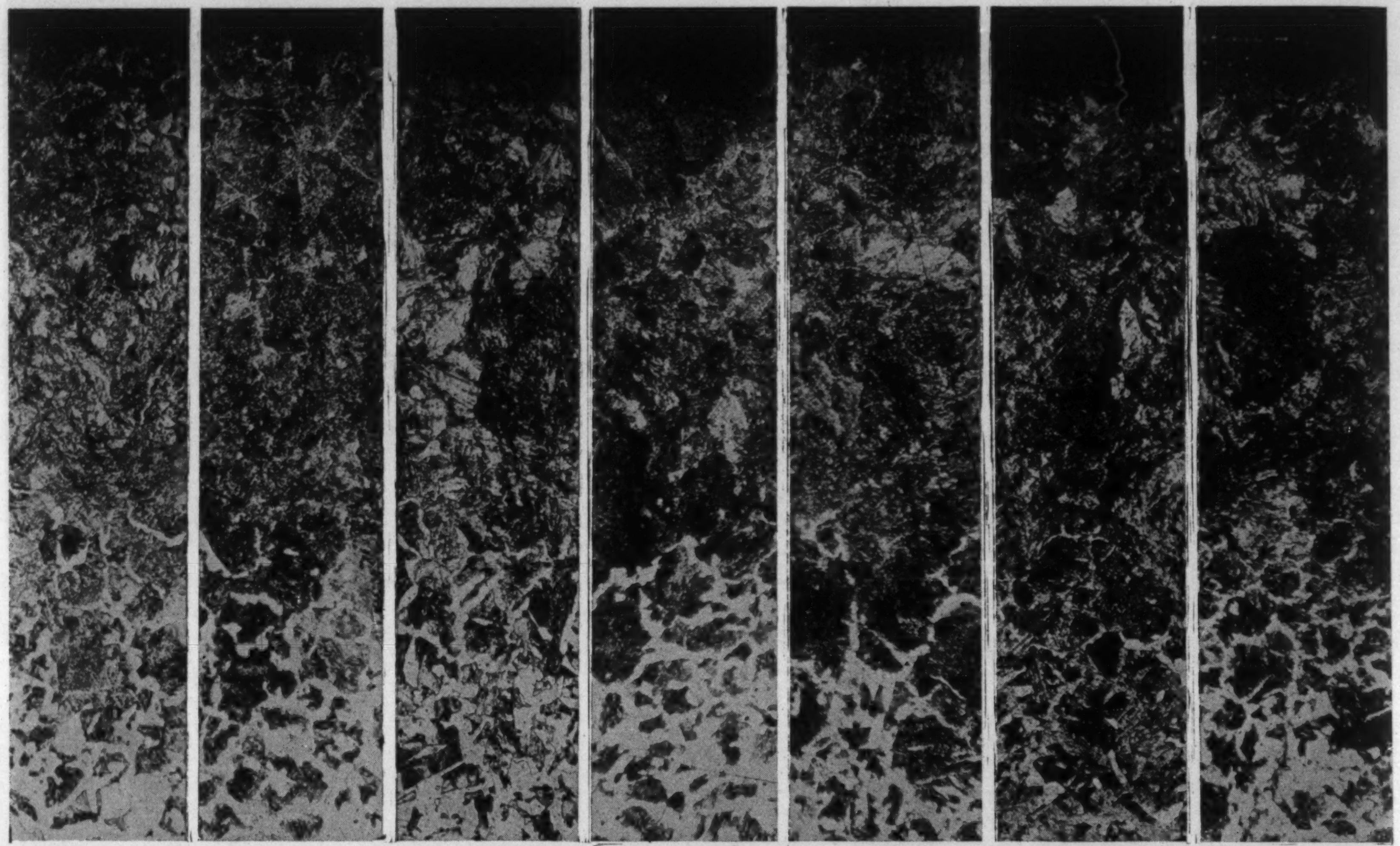
A water gage indicates the muffle pressure and although 3″ is easily obtainable, the operating pressure is only ¾″ which produces equally good results.

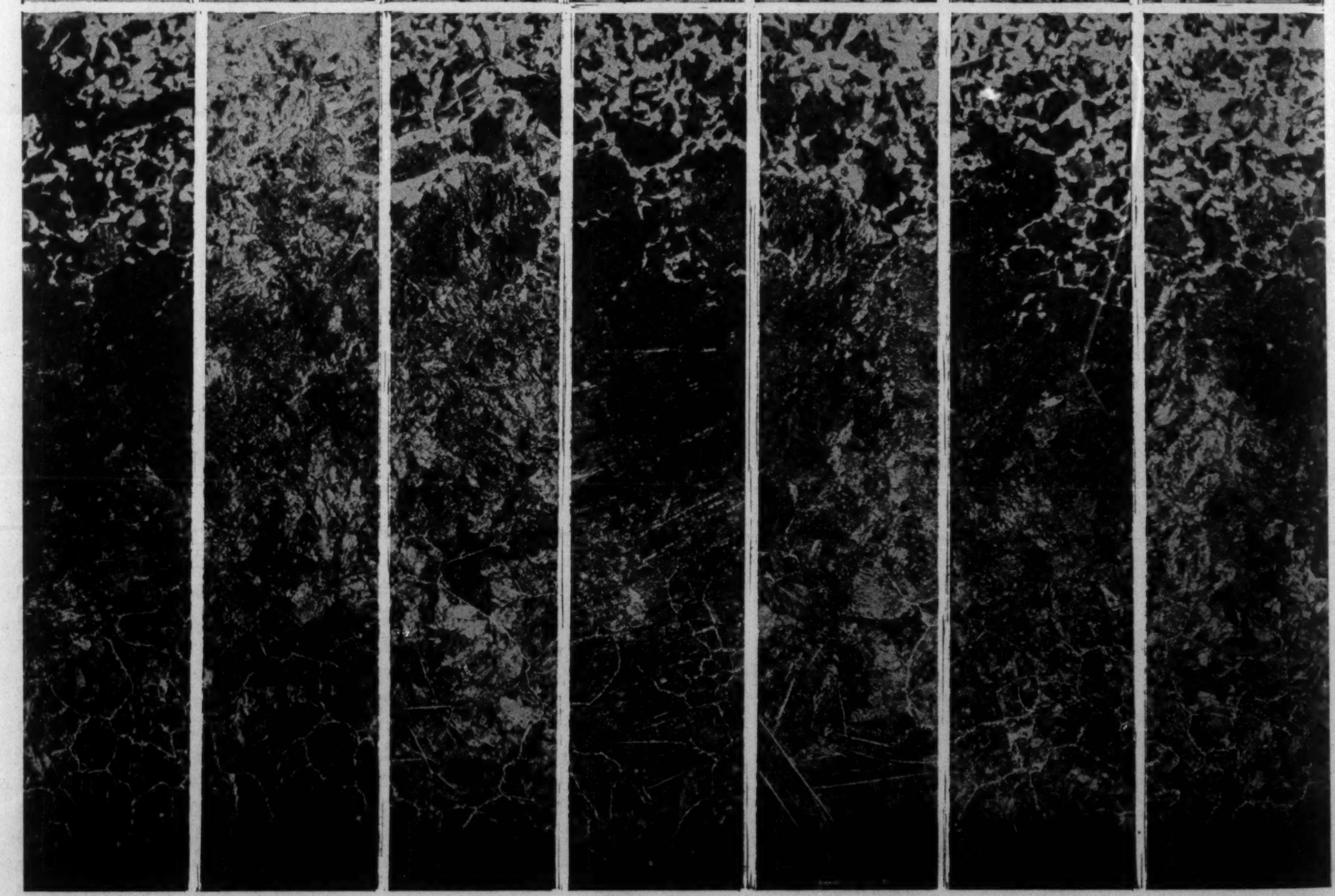
The panel board shown on the facing page contains natural and flue gas differential draft gages on which constant pressures of both gases are maintained. It also contains a tube in which exhaust muffle gas burns at all times. A record of the temperature of this flame is obtained by means of a couple inserted in the tube and the repeating cycle shown on the chart is indicative of both the composition of the exhaust gases and the proper functioning of the furnace and its operator. Supple-

TEST FOR UNIFORMITY OF CARBURIZATION ACROSS TRAY

All samples finished all over from same bar SAE 1020 Steel. Magnification 100 ×. C 0.15-0.25, Mn 0.30-0.60, P 0.045 max., S 0.05 max.

Met. No. 2629 Met. No. 2631 Met. No. 2633 Met. No. 2635 Met. No. 2637 Met. No. 2639 Met. No. 2641



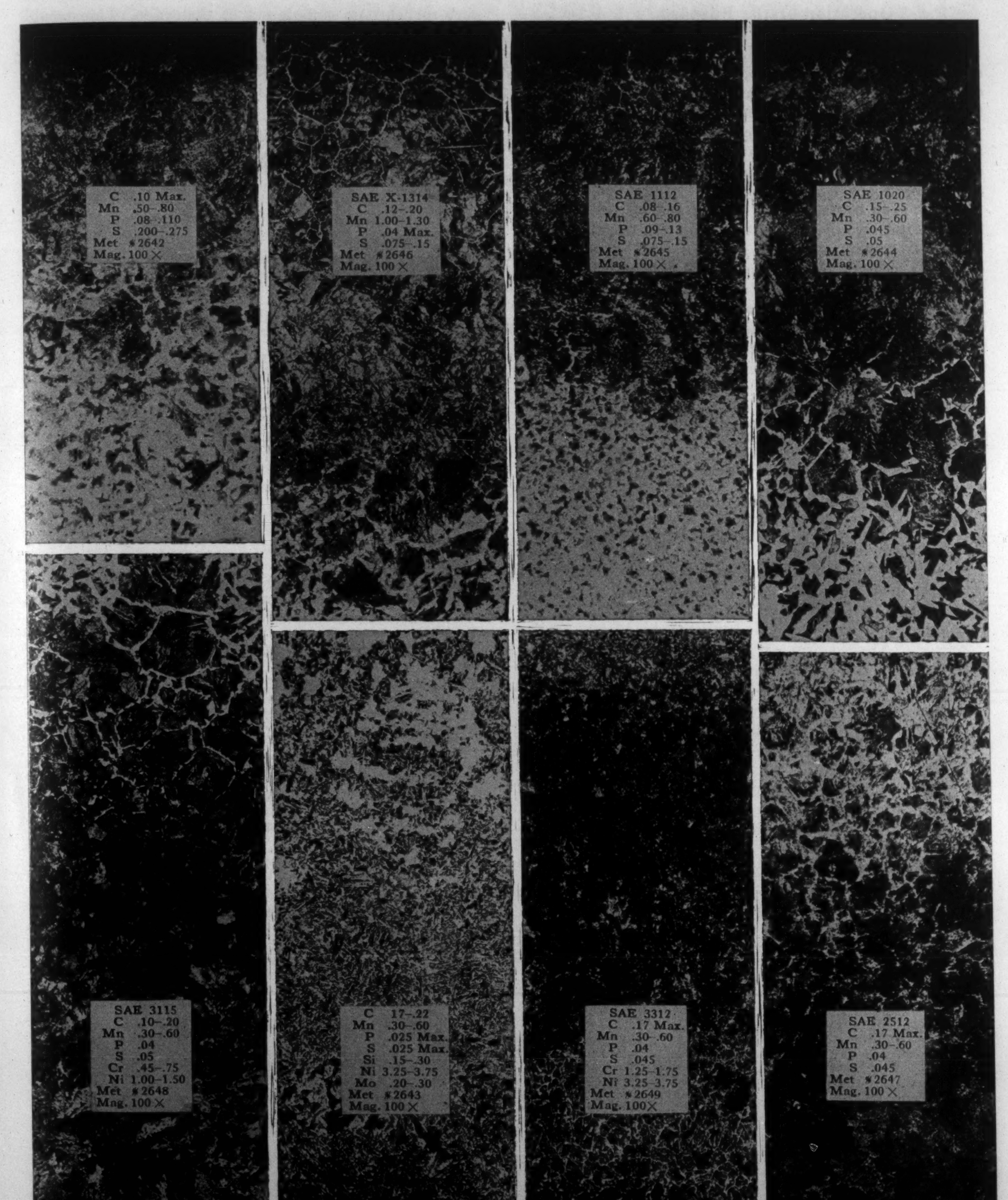


Met. No. 2628 Met. No. 2630 Met. No. 2632 Met. No. 2634 Met. No. 2636 Met. No. 2638 Met. No. 2640

All samples finished all over from same bar SAE X-1314 Steel. Magnification 100 ×. C 0.12-0.20, Mn 1.00-1.30, P 0.04 max., S 0.075-0.19.

VARIOUS TYPES CARBURIZING STEELS

All Carburized at the Same Time.


C .10 Max.
Mn .50-.80
P .08-.110
S .200-.275
Met #2642
Mag. 100 ×
SAE X-1314
C .12-.20
Mn 1.00-1.30
P .04 Max.
S .075-.15
Met #2646
Mag. 100 ×
SAE 1112
C .08-.16
Mn .60-.80
P .09-.13
S .075-.15
Met #2645
Mag. 100 ×
SAE 1020
C .15-.25
Mn .30-.60
P .045
S .05
Met #2644
Mag. 100 ×
SAE 3115
C .10-.20
Mn .30-.60
P .04
S .05
Cr .45-.75
Ni 1.00-1.50
Met #2648
Mag. 100 ×
C 17-.22
Mn .30-.60
P .025 Max.
S .025 Max.
Si .15-.30
Ni 3.25-3.75
Mo .20-.30
Met #2643
Mag. 100 ×
SAE 3312
C .17 Max.
Mn .30-.60
P .04
S .045
Cr 1.25-1.75
Ni 3.25-3.75
Met #2649
Mag. 100×
SAE 2512
C .17 Max.
Mn .30-.60
P .04
S .045
Met #2647
Mag. 100 ×

mentary gas analyses are made, however, as a positive check on this chart. It is considered unsafe to operate with a CO percentage of less than 5% in the exhaust gas because of improper carburizing results.

Operation of the Furnace

Regardless of production conditions, the furnace is never cooled below 1400° F. and when not in use requires practically no attention. However, after a period of inaction, and before any natural gas is introduced into the muffle, it is essential to remove all O_2. This is accomplished by purging with flue gas only for a period of about one-half hour. A gas analysis is then made to insure the absence of all O_2 and the furnace *is ready* for operation.

Trays, each containing 35 worms and 35 sectors, are then charged into the muffle at 18 minute intervals. Since the muffle accommodates 22 trays, a total time of 6.6 hours is necessary to produce 0.035″ case on the sector and 0.040″ on the worm. The discrepancy in case depths is due to the difference in analysis of materials used for these parts. The X-1314 steel used in the worms carburizes correspondingly faster than the straight carbon type of which the sector is composed.

When the time clock rings an 18 minute interval, the following sequence of operations is performed:

1. The muffle discharge valve raises, a tray is drawn into the holding chamber and the valve closes.
2. The discharge door of the holding chamber is opened, the work is quenched in oil, the tray is returned by a roller conveyor and the door is closed.
3. The charge valve raises and the automatic pusher moves another tray into the preheat section of the muffle.
4. The charge valve closes and another tray is placed into the purging vestibule.

All of these operations are controlled by the operator at the discharge end of the furnace. Several minutes before the next push, another tray must be rolled into the purging chamber and the cycle is repeated.

Discussion of Results

All worms and sectors receive a 100% Rockwell hardness inspection, a low limit of C-60 being specified, and occasional tell-tales are broken daily to check case depths. The hardness on both parts has been extremely uniform, averaging C-63—C-64 after water quenching from a tube furnace and drawing at 325° F. in a mesh belt conveyor furnace, heated exclusively by the waste gases from the hardening furnace. Rejections for hardness are very rare. No rejections have as yet been made for case depth or character.

The elimination of distortion is also very important on these parts and has been accomplished to an unexpected degree. A uniform shrinkage in the hole and overall length of the worm of about 0.003″ has been found to occur and must consequently be considered in the machining of the part. No appreciable change is found in the sector. The photograph on this page shows an etched longtitudinal section of a hardened worm and illustrates distortional possibilities.

A number of experimental runs have been made on various production parts which have long been the subject for chronic complaints of distortion. This trouble has been easily traceable in each instance and has been found to result from any one or several of the following conditions:

1. Faulty packing of parts in carburizing box.
2. Impossibility of uniform heating of parts in a carburizing box.
3. Effect at high heats of weight of other parts and carburizing compound on pieces in the lower section of the box.
4. Shifting of parts as the boxes are pushed through the furnaces.
5. Dumping boxes while parts are still too hot.

All possibility of distortion emanating from the conditions enumerated above is eliminated by the use of this continuous process and every experiment has shown excellent results. For example, the average free-wheeling drive shaft has a run-out on the cup of approximately 0.004″ after box-carburizing, with occasional instances of 0.010-0.015″. Experimental runs in the gas furnace indicate the possibility of maintaining a run-out limit of as low as 0.003″, even after hardening. The case depth specified on this part is 0.065″-0.075″. Similar results have been found on gears and pinions. The necessity for straightening on parts which permit this operation can also be materially reduced.

Summary

A discussion of the factors governing the adoption of this process and equipment may now be easily understood.

1. First cost.

The first cost compares very favorably with that of a well-designed gas-fired box carburizing furnace of the same capacity.

2. Economy and simplicity of operation.

The total cost of all natural gas used for both heating and carburizing is approximately $0.001 per pound of work. Total gas consumption per hour is about 725 cubic feet. All other charges, including labor, power, depreciation, etc., raise this total to $0.0031 per pound. This figure is extremely low. The operation of the furnace is so simple that less than one-half of one man's time is required.

3. Precautions.

The only precaution required to insure absolute cleanliness is the provision for the removal of carbon-laden gases resulting from high pressure air-cleaning the hot valve seat. This operation is performed every 30 days and is completed in about 5 minutes. A hood, suitably located above the discharge door removes these obnoxious gases.

4. Uniformity of case depth and carbon concentration.

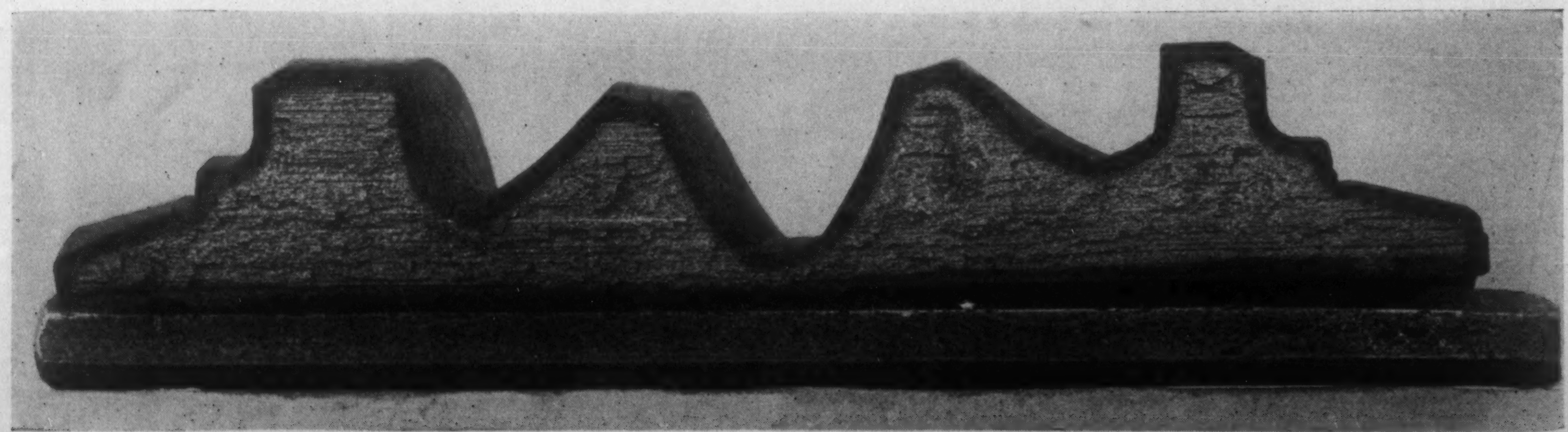
The proper manipulation of time, temperature and gas ratios insures a high degree of uniformity.

5. Rate of carbon penetration.

The rate of penetration in the actual carburizing zone is high and even at 1650° F. shows a minimum of approximately 0.010 per hour on any standard type of carburizing steel.

6. Facility for direct quenching.

This equipment is obviously ideal for direct quenching.



Fracture of Worm. Etched to Show Uniformity of Case. SAE X1354. Mag. 2 ×.

Effect of Controlled Cooling and Temperature Equalization on Internal Fissures in Rails

By C. P., O. F. A. and N. P. P. SANDBERG

ONE of the most serious and dangerous causes of the failure of rails is the "internal fissure." As axle loads have increased and the hardness of the rails brought up to withstand them by increasing the hardening ingredients in the steel, the percentage of these failures has become more noticeable. Statistics indicate that certain heats of steel are more liable to fail from this cause than others and it is probable that hot bed conditions which lead to any rapid chilling of the rails when cooling are important factors. It may be taken, in fact, that different heats of steel have a varying intrinsic susceptibility to this type of defect which may be favored or restrained by conditions in the mill, and we know that heats containing high carbon and other hardening ingredients are as a rule more susceptible than softer steels.

IN connection with the subject of this article, two recent papers by G. W. Quick of the Bureau of Standards will be found of interest. These are "The Resistance to Impact of Rail Steels at Elevated Temperatures," and "Tensile Properties of Rail Steels at Elevated Temperatures" presented at the February 1932 meeting of the American Institute of Mining and Metallurgical Engineers. The article also appears in Bureau of Standards Journal of Research, Vol. 8, 1932, pages 191-225 (Research Paper No. 409).

It is now generally accepted that rails at the mill and before they are laid in the track may contain small internal ruptures lying at or near the center of the head, and that there is a possibility that such ruptures may be the starting points of the failures. Similar defects have been found in other types of steel forgings—especially alloy steels. In rails the defect has been referred to as "shattered zone."

The ruptures in rails before being laid in the track are usually extremely fine (*i. e.,* the separation of the two faces is almost ultramicroscopic) and are characterized by showing an approximately circular area of distinctive texture when broken through. Their presence may be discovered by deep etching of sections or by examination of polished sections by the magnetic "iron dust" method, but no method of external examination without sectioning or breaking has yet been devised to indicate their existence until opened out and extended by subsequent stresses in service. The liability to form these cracks undoubtedly increases under certain cooling conditions, and it was particularly since their avoidance became a very necessary precaution in heat treated rails that the authors made an intensive study of the problem which has resulted in the introduction of the "Sandberg 'Oven' Process."

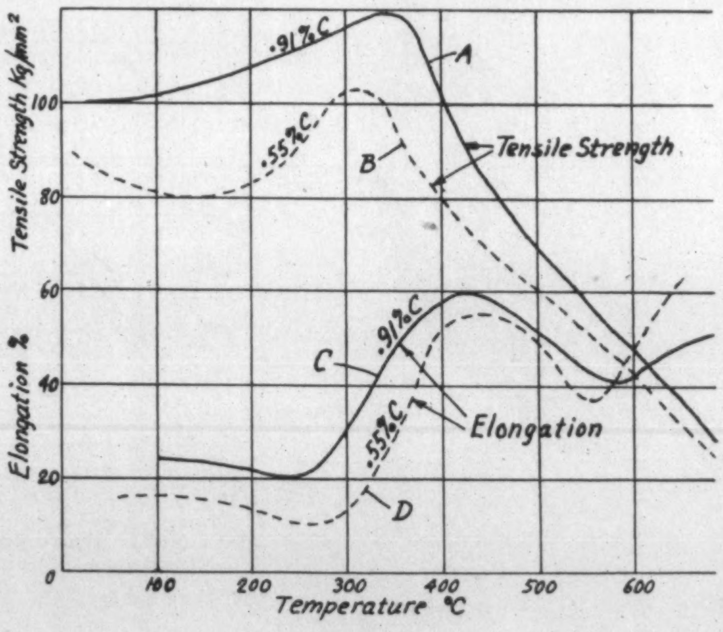


Fig. 1

This process has now been in practical operation in England for over two years and it has been proved to form a practical safeguard against the formation of internal ruptures even in steel which is particularly susceptible to their formation. Before describing the practical applications of the process the following experimental and theoretical basis on which it is founded may be of interest:

INTERNAL STRAINS SET UP DURING THE COOLING OF RAILS

In considering the formation of stresses during cooling we have three important factors to take into consideration, namely:

1. Strength
2. Temperature differences
3. Rates of contraction

1. Strength

This subject has received a considerable amount of study, and the general variations in the strength of a steel of known composition when cooled from a high temperature are accurately known. Fig. 1 shows curves representing the tensile strength and ductility of two steels—(0.91—0.55% C. respectively) which were obtained by Depuy and published in the Journal of the Iron and Steel Institute.[1]

Starting from a temperature of about 700° C. it will be seen that the tensile strength of the steel increases very rapidly down to a temperature of about 350° C. and then gradually falls away to 250° C. and rises again towards atmospheric temperature. The elongation has a very definite maximum just over 400° C. and very definite minima between 500° and 600° C. and between 200° and 300° C.

A number of similar results were obtained for rail and other steels by Freeman and Quick[2] which in general follow Dupuy's results. It was, however, noted especially by them that the minima of ductility (both elongation and reduction of area between 400° and 600° C.) is very much more marked in certain heats than in others.

In connection with their investigation into the formation of internal fissures in rail steel the authors of this article carried out some tests by means of the slow bending of notched bars by the Humfrey slow bending machine and the results are shown in Fig. 2. They confirm the tensile tests by other investigators, and again bring out very clearly the minimum ductility in the vicinity of 500° C. and the very rapid increase of strength in cooling between 600° and 350° C.

Two important points which are not brought out by the above strength-temperature curves must also be considered:—

(a) At temperatures above 400° C. the rate at which steel is strained has a very marked effect on its tensile strength, and a less marked though probably equally important effect on its ductility. At slow rates of strain the tensile strength is very greatly reduced and the steel behaves under stress more as a very viscous fluid than as a solid. It is probable (though results obtained are not conclusive) that the ductility also is diminished under very slow strains.

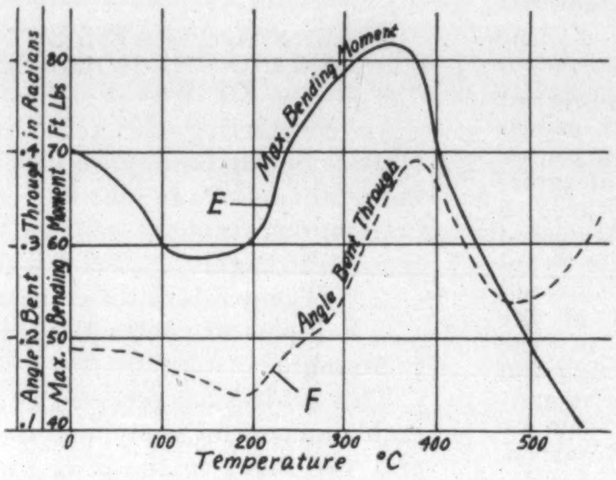


Fig. 2

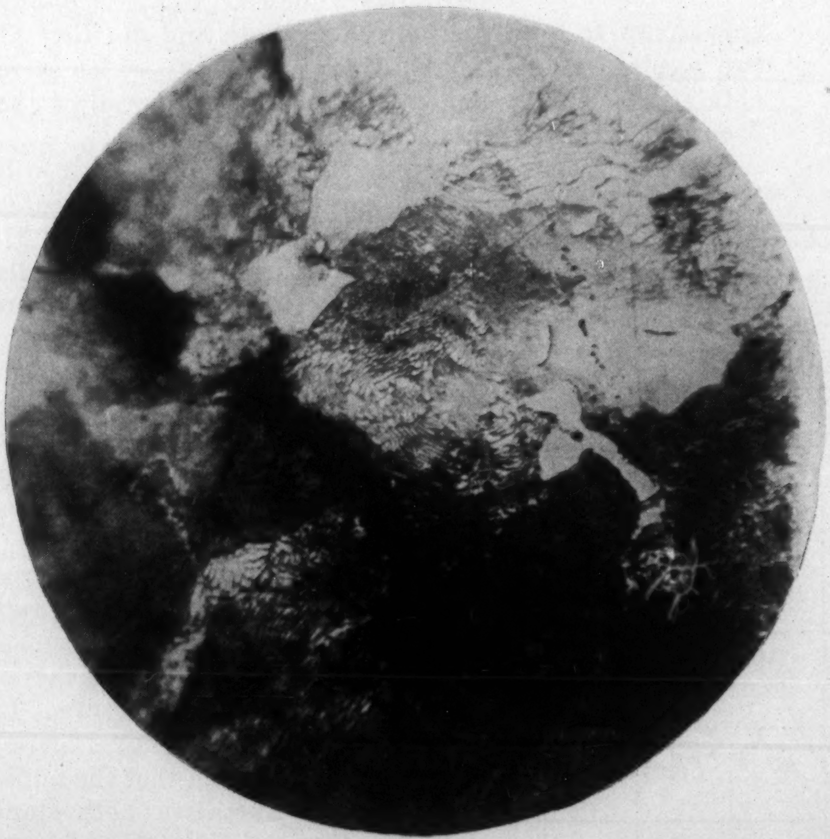


Fig. 3. Crack in bend test piece broken at 440° C. Mag. × 1000.

(b) While at normal atmospheric temperature and up to moderately elevated temperatures the mechanism of strain takes place by individual distortion of the crystals themselves by slip along their gliding planes, at temperatures above this the mechanism of strain begins to alter and the crystals themselves tend to move bodily over one another. If the strain is prolonged to the point of fracture then at normal atmospheric and up to moderately elevated temperatures the line of fracture will be found to pass through the crystals while above this it tends to pass round them. The later phenomenon was especially noted by the authors in their notched bar tests and was confirmed by Freeman and Quick. A photomicrograph of a crack found near the fracture of a bend test piece broken at 440° C. illustrates this (Fig. 3). It is characteristic of shatter cracks found in rails and steel forgings that the path of the cracks follows very definitely round and not through the crystals.

2. Temperature differences

It is an impossibility to cool any mass of steel from a high temperature without setting up temperature differences between one part of a mass and another. The slower the average rate of cooling throughout the mass, the less these temperature differences will be, but they can never be absolutely avoided. Temperature differences may be set up owing to two causes: (a) lack of uniformity of section and (b) conduction gradients.

(a) In an irregular section such as a rail allowed to cool naturally, the thinner parts such as the web and flange must always cool at a quicker rate than the more massive head, and as shown by Freeman and Quick in the latter part of their paper, differences of more than 100° C. may occur even in a slowly cooled section.

(b) *Conduction gradients.* The temperature gradient between the inside and outside of a mass depends entirely upon the rate at which heat is withdrawn from the surface. In a mass comparable in dimensions to the head of a rail it has been shown that even when cooling in still air temperature differences of up to 50° may be set up between the center and outside of the head. These temperature differences refer to experiments on a rail reheated to a uniform maximum temperature of about 900° C., but in the ordinary rolling of rails the outside portions of the section are very seriously chilled by contact with the rolls, water on the rolls, etc., so that temperature differences between the outside and the inside may be present to a much greater extent. If during the early periods of cooling certain parts are losing heat faster than others a reversal of rate of cooling must

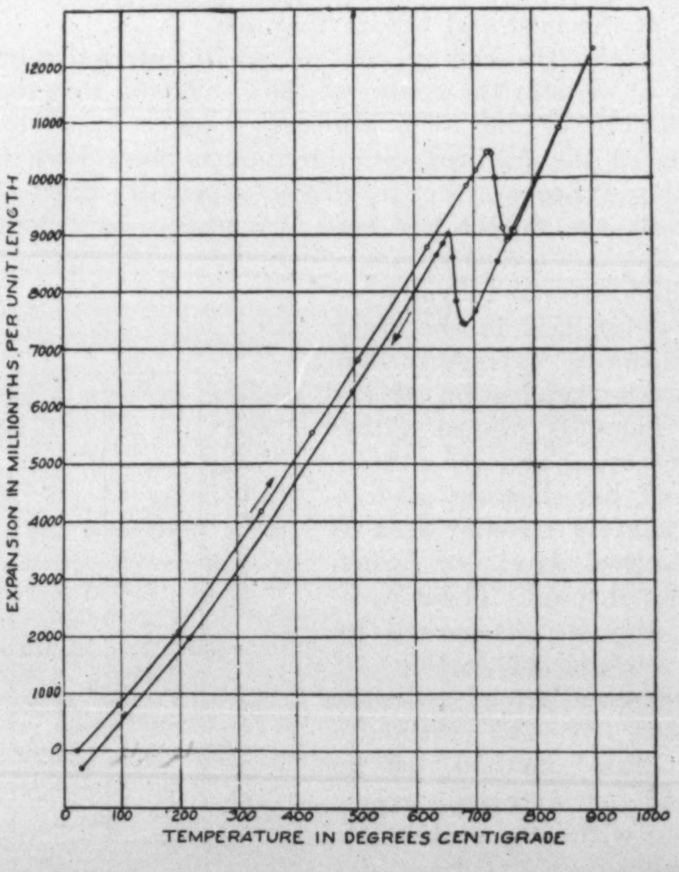


Fig. 4. Thermal expansion curve of steel of rail composition. C, 0.59; P, 0.024; Mn, 0.92; S, 0.033; Si, 0.25. (U. S. Bureau of Standards, Scientific Publication, No. 433.)

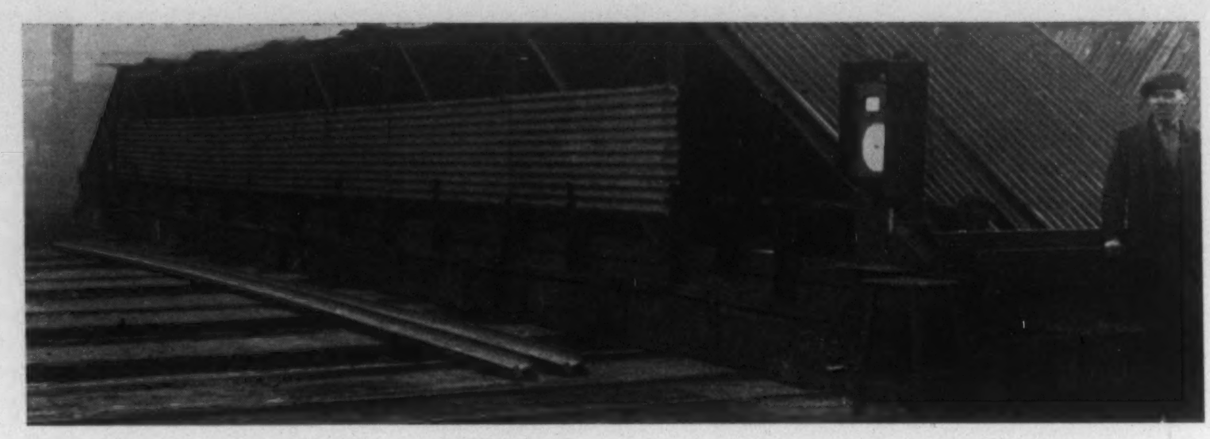


Fig. 5. Sandberg Oven at works of Cargo Fleet Iron Co., Ltd. (Rail entrance side.)

occur at some later stage and portions which were originally cooling slower must subsequently cool quicker.

3. Rates of contraction

The rate of contraction of steel is, like its specific heat, subject to sudden changes when passing through the critical range. A uniform rate of contraction occurs above the critical range, a sudden expansion during the critical range and a steady contraction from this temperature to atmospheric. (See Fig. 4 as shown by Freeman and Quick on page 43 of their paper.) This sudden expansion at the critical temperature is, similarly to the specific heat, dependent upon the rate of cooling and may be partially suppressed by rapid cooling. At the same time the final size obtained at atmospheric temperature varies also with the rate of cooling, a rapid cooling increasing the ultimate specific volume of the steel.

COOLING OF RAILS

The rails will presumably leave the rolls at a temperature of about 950° C. and, owing to their section, the flange and web will cool much more rapidly than the head. While the whole rail is still above the critical temperature, the contraction of these parts will therefore be quicker and the rail will tend to camber with the flange concave. As soon, however, as the flange and web reach the critical temperature, a sudden expansion occurs in them and the camber is rapidly reversed so that the flange becomes convex and the head concave. At a still later stage the head passes through the critical range and in turn undergoes sudden expansion so that the previous curvature is reversed and the head again becomes convex and the flange concave. From this point down to atmospheric temperature this camber is gradually reduced, but owing to the fact that the web and flange have been cooled through the critical range at a quicker rate than the head, their final length will be greater and therefore before cold a further reverse camber takes place and the rail has the flange convex and the head concave.

To avoid this curvature of the rail when cold it is of course a practice to put an opposite camber in the rail immediately after leaving the rolls so that it finally cools down approximately straight.

During certain stages of this cooling when web, flange and outside of the head have become fairly cold and therefore their rate of contraction small, the interior of the head will be at a higher temperature and owing to conduction to the cooler outside will be contracting at a more rapid rate than these cooler layers.

We see that at a certain stage of cooling the interior of the head must be contracting at a faster rate than the remainder of the section and thus be subject to tensional strains which may not be only tensional strains in one direction but tensional strains in all directions equivalent to the negative of a fluid pressure. At what point in the cooling are such tensile strains likely to be most dangerous and to lead to actual cracks forming in the interior of the head?

Returning again to the strength temperature diagram shown above (Figs. 1 and 2), consideration indicates that the most dangerous part of the cooling range is that between the critical range and about 350° C. During certain stages of this period the steel possesses very low tensile strength and as shown on the diagrams has a marked minimum of ductility. At 350°—400° C. steel has its maximum tensile strength and therefore if the exterior portions have fallen to or below this temperature while the interior portions are still at substantially higher temperature then the rigidity of the outer layers can exercise their maximum effect in restraining the interior from contracting as it cools. In addition, the evolution of heat which occurs during the critical range within the center of the head leads afterwards to an increase in the temperature gradient and therefore to a period of more rapid cooling of the interior portions by conduction.

Fig. 6. End View of Sandberg Oven at Cargo Fleet Iron Co., Ltd.

These conclusions were arrived at by the authors after many years' study of the problem of the cooling of rails and of their heat treatment, and the "Sandberg Oven Process," for which patent protection was applied for in 1928, was arrived at as a direct result of these conclusions. It was therefore of very great interest when Freeman and Quick published their paper in which these conclusions as to the formation of internal fissures in rails during cooling were so fully supported. Freeman and Quick state in their paper: "The data indicate that equalization of temperature and slow cooling of rails and some other steel products through their secondary brittle range are desirable procedures to follow in order to prevent formation of shatter cracks."

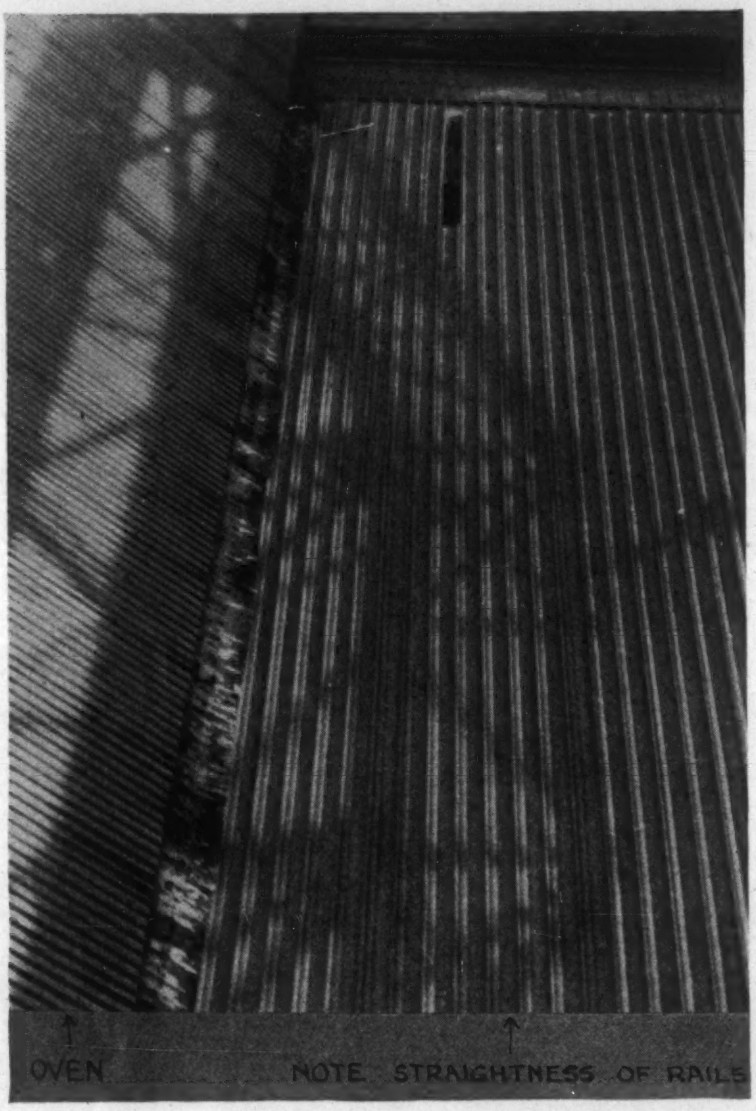


Fig. 7. Rails which have been treated by the Sandberg Oven.

From the above it has been shown that it is impossible at some stage of the cooling to avoid tensile strains being set up in the center of the head, and the problem of avoiding any damage from these strains is therefore one of insuring that the tension shall come at a time when it can do no harm and that its intensity shall be a minimum.

The essentials of the "Oven" process are as follows:—

Before the major portion of the rail metal, particularly the metal of the rail head, has cooled below a temperature of about 500° C., the whole rail is placed in a protected, and, if necessary, preheated chamber which substantially prevents further radiation or conduction of heat from the surface and allows the whole section to attain a substantially uniform temperature before further cooling proceeds.

After this equalization, partially retarded cooling may be considered desirable in order to insure that no substantial strains are set up between the inside and the outside during the cooling to a temperature of about 350° C. If the rail has been thoroughly evened out, say at a temperature of 500° C., such strains will hardly arise even with normal cooling.

The "Oven" is a cover, or chamber, in which the rails are sheltered for a short period, preferably soon after they have cooled through their critical range, and while their external temperature is still well above the maximum tensile strength temperature of about 400°—350° C. As soon as the rails are pushed into the "Oven"—which is usually kept at a temperature of about 400° to 500° C.—a check takes place upon any further cooling of their exterior, and accordingly the higher temperature of their interior levels up, or "evens out" the temperature of the whole rail, with the result that in a few minutes there is no appreciable temperature difference and consequently no stresses are generated.

Fig. 5 shows an "Oven" as installed by the Cargo Fleet Iron Co., Ltd., of Middlesbrough, England, who have been the pioneers in adopting the process. The two rails on the hotbed are ready to be pushed into the "Oven" under the vertical swinging doors, and a recording pyrometer may also be seen, by means of which the temperatures are noted, and any desired control made. Fig. 6 shows an end view of the same "Oven," the left-hand portion of which is that into which the rails enter. Fig. 7 shows a typical example of the straightness of rails which have been through the "Oven." This "Oven" is built in two portions, the front one of which is lined with a refractory material and fitted with gas burners to heat it up when necessary. It is in this first portion that the equalizing of temperature of the rails takes place. The second portion consists of a shelter in which partially retarded cooling of the rails takes place after they have been "evened out" in the first portion, and from which after having cooled to about 350° C. they finally pass out onto the hot bed where their cooling out proceeds normally. The "Oven" stands across the full width of the hot bed, in order to be able to take the longest rails rolled, and has recently dealt most satisfactorily with rails of 90 feet in length.

Since the installation of this plant there has been a notable fall in the percentage of rail breakage at the straightening presses, coupled with a minimum of test failures, and a reduction in the amount of cold-straightening.

Five of the leading British railmakers have now installed similar "Ovens," and there is every prospect of the process becoming general practice in England for all rails. About 80,000 tons of rails have already been treated in these "Ovens," and in many cases contracts now being placed specify "Oven" treatment. Great interest is now being shown in the "Oven" process by the railways and rail mills on the continent of Europe. In the United States the Bethlehem Steel Company will very shortly have one of these "Ovens" at work at their Lackawanna Rail Mill, by means of which they will be able to demonstrate the advantages of this process for American conditions.

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[1] E. L. Dupuy. Journal Iron and Steel Institute, Vol. 104, Part 2, 1921, pages 91-116.

[2] American Institute of Mining and Metallurgical Engineers Technical Publication No. 269, 1930; Extended abstract in Metals & Alloys, Vol. 1, June 1930, pages 556-558.

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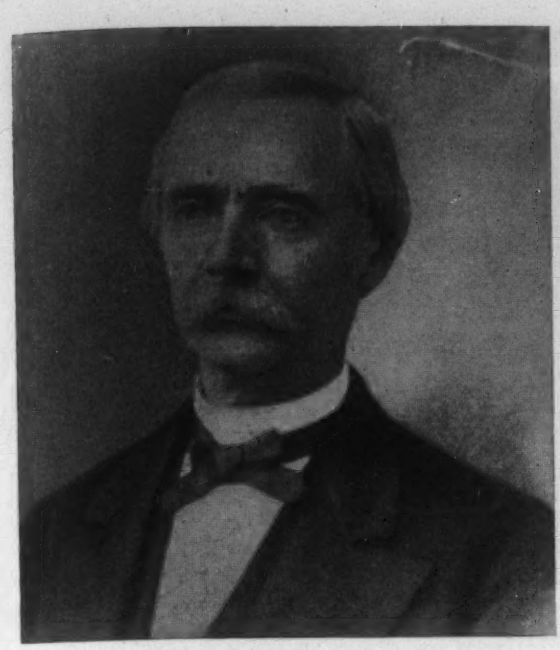
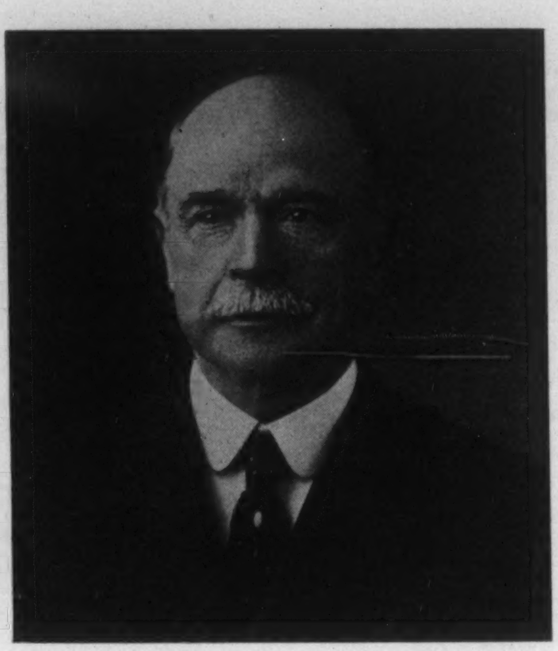
Fred Archer, formerly president of Fred Archer, Inc., has been appointed to the Philadelphia office of the Lincoln Electric Company. He is in charge of motor sales for this office. Archer was with the White Motor Company and the Kelly-Springfield Tire Company before forming his own concern.


William Kelly—the American Inventor of Bessemer Steel.


Captain Robert W. Hunt, the First Works Chemist, Cambria Iron Co., 1860.

Daniel J. Morrell — Manager of old Cambria Iron Co., who "backed" Kelly.

EARLY CHEMISTRY and CHEMISTS in the IRON and STEEL INDUSTRIES. No. 2

The Cambria Iron Company, The First Steel Chemist and Beginnings of Bessemer Steel

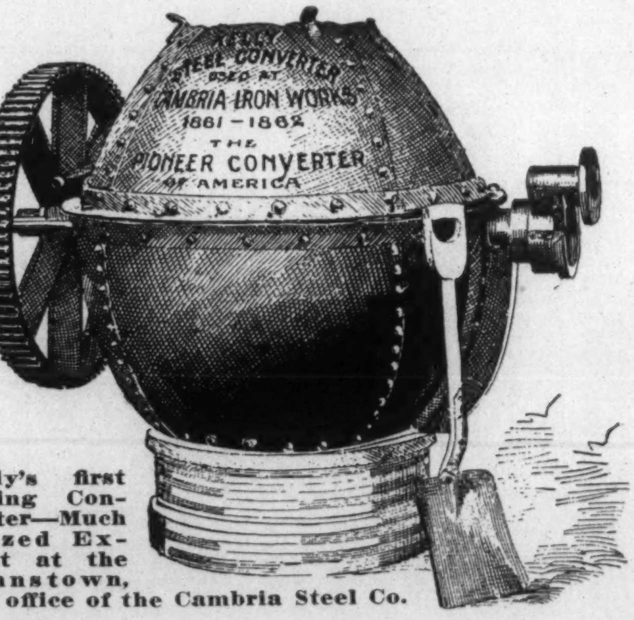
L. W. SPRING and L. E. GILMORE

Early Days in Iron and Steel

THE Cambria Iron Company and some of the men connected with it played a large part in early experimentation with and in establishing the Bessemer process in this country. Although Cambria did not build her own Bessemer plant until 1871, the sixth such plant in the United States, nevertheless the first tilting steel converter in the United States, ½ ton capacity, was built at Cambria in 1861 by Wm. Kelly, the American inventor of Bessemer Steel. This converter is still a prized possession of the Cambria works at Johnstown, Pa.

Andrew Carnegie, writing for the New York Evening Post Jan. 12, 1901, states:

> "From 1831 until as late as 1860 little progress was made in developing the manufacture of steel, for the total product in Pennsylvania in 1850 was only 6,000 tons, still principally blister steel ....It was not till 1864, ... that the revolution in steel manufacture came to us, and the Iron Age began to give way to the new King Steel, for our first Bessemer steel was made in that notable year."



Kelly's first Tilting Converter—Much Prized Exhibit at the Johnstown, Pa., office of the Cambria Steel Co.

With the inception of the great Bessemer steel industry, which for many years was to play such a commanding part in the life of the nation, began the rise of the steel chemist. Perhaps a main reason for the need of a chemist in the new Bessemer process was the speed of the converter process which completed a heat of steel often of five or more tons in about 20 minutes. This, of course, was revolutionary compared with the former crucible process which required hours for the completion of a crucible of steel of perhaps one hundred or more pounds.

The expensive crucible steel was used for tools, largely. Much wrought iron was made from pig iron by the puddling process. This latter, of course, was the soft iron which the new Bessemer steel to a considerable extent replaced. The prolific product of the Bessemer converter brought about the vast growth in the use of soft steel as structural material. The wrought iron puddling process, of course, is slow and expensive compared with the Bessemer process since two men make only 600 pounds in 2 hours, perhaps, instead of the 20 minute completion of 5 to 20 tons mentioned above.

William Kelly and His Invention

The Bessemer process played such an important part in our civilization[1] and also in establishing the earlier steel chemists, that a sketch of the inception of this process seems in order.

In the woods, near Eddyville, Kentucky, the town where a certain William Kelly established himself following a chance visit to the graduation exercises of a girl's school where a comely maiden, Mildred A. Gracey, graduated, the first steel converters were experimented with during the years 1846 to 1857. Kelly, with his brother, purchased ore lands here. Cutting wood for charcoal, they made wrought iron by the "Finery Fire" process. The product, being made into kettles, plowshares, bars, general utensils and implements at "forge fires," was sold up and down the river and in the neighborhood.

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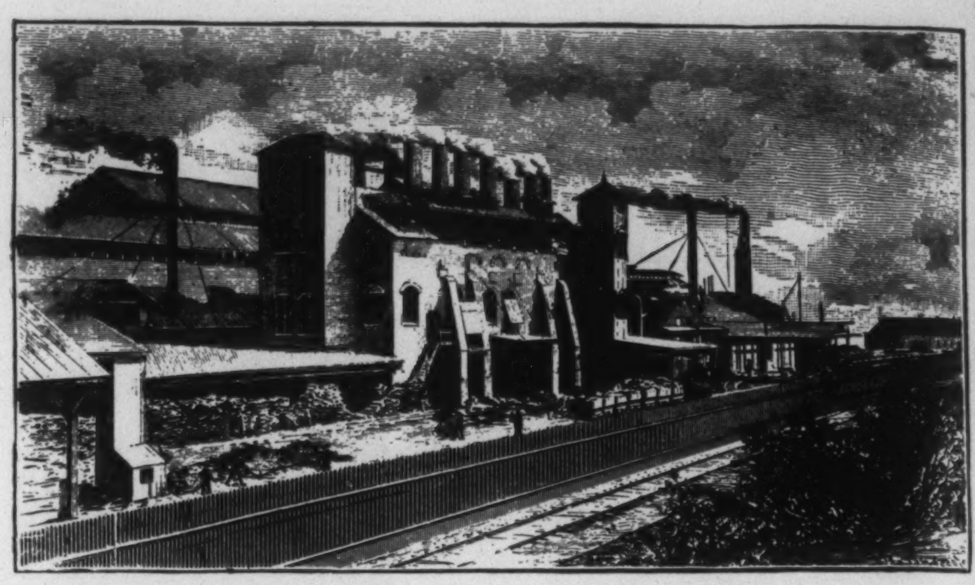
[1] "The Romance of Iron and Steel—The Story of a Thousand Millionaires." H. N. Casson. *Munsey's Magazine*, April, 1906, and on; in book form, A. S. Barnes & Co., New York.



Bessemer Steel Works of Troy Steel & Iron Co. (Illustrated from "Troy and Vicinity" by Arthur James Weise, published in Troy, 1886.)

It is related that, searching for some easier method, Kelly noticed that by blowing a stream of air directly on the *molten* metal it glowed and appeared to become hotter. Proceeding from this basis he soon found that his iron could be made in considerably less time. However, his customers learned that he was changing the process somewhat and insisted that they must have iron produced by no "new-fangled" but by the old process. Kelly, however, proceeded with his experiments, though relatives and neighbors considered him "crack-brained," building apparatus after apparatus in an adjacent woods where he would not be interfered with. Here, during several years, he experimented. However, having insufficient capital he failed to make his process a really commercial one.

Now Daniel J. Morrell was the very energetic manager of the Cambria Iron Company of Johnstown, Pennsylvania. When Kelly practically failed of funds to go ahead with his experiments, Morrell gave him a place in the Cambria plant and financed a small converter which Kelly built of a tilting instead of his former stationary type. After failure, amid jeers, in the first test, Kelly succeeded in the second one, though full development on a commercial scale still was some distance away.

This process, locally, had been known as "Kelly's Air Boiling Process" and the company formed to exploit it was called "The Kelly Pneumatic Process Co."

Thus, the Bessemer process, for which credit usually goes to Sir Henry Bessemer of England, was invented independently in this country by William Kelly and by Bessemer in England. There are stories and claims that Bessemer had visited Eddyville and learned of Kelly's process, before his own was developed,[2] but certain it is, that Bessemer put the process on its big financial and commercial basis, and it seems probable that both men invented the process independently of each other.

Kelly had been granted patents in 1857 for his pneumatic process by the United States Patent Office. Although Bessemer had been granted patents in 1856, Kelly's claims to priority were recognized by the patent office and his patent granted, though applied for in 1857. Upon expiration of these patents, Kelly's patent was renewed by the United States Patent Office, while refusal was made to Bessemer's application for renewal.

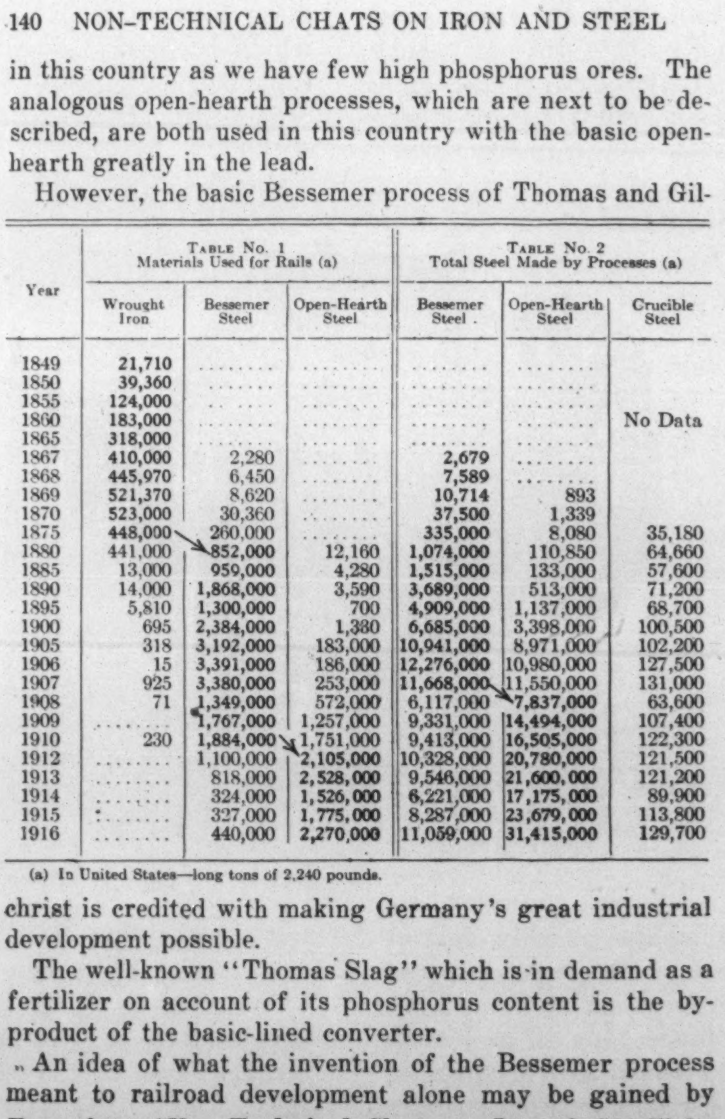
First Bessemer Steel in the United States Was Made by the Kelly Pneumatic Process

The Kelly Pneumatic Process Co. was formed by Daniel J. Morrell of Johnstown, E. B. Ward of Detroit, William M. Lyon and James Park, Jr. of Pittsburgh and Z. S. Durfee of New Bedford, in 1863, and later (in 1865), Charles P. Chouteau, James Harrison and Felix Valle—all of St. Louis—joined the company. Z. S. Durfee went to England and obtained the rights to use Mushet's patents on spiegeleisen and recarburization, without which both Bessemer's and Kelly's processes would have been failures.

The Kelly Pneumatic Process Co. built an experimental Bessemer plant in Capt. Ward's extensive iron works at Wyandotte, Michigan.

It is said in Chronology of Iron & Steel by Goodale and Speer that Wm. F. Durfee established a well equipped laboratory at the Wyandotte plant. Probably this was patterned after the first known iron and steel works laboratory established at Cambria Iron Company by Capt. R. W. Hunt in 1860.

It was at the Wyandotte plant in the fall of 1864 that William F. Durfee, a cousin of Z. S. Durfee, made the first successful "blow" of Kelly's product (now so well known as Bessemer Steel) produced in the United States. The ingots cast from this steel were shipped to the North Chicago Rolling Mills of Chicago, where on May 24, 1865, they were rolled into the first steel rails produced in this



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in this country as we have few high phosphorus ores. The analogous open-hearth processes, which are next to be described, are both used in this country with the basic open-hearth greatly in the lead.

However, the basic Bessemer process of Thomas and Gil-

Year	Table No. 1 Materials Used for Rails (a)			Table No. 2 Total Steel Made by Processes (a)		
	Wrought Iron	Bessemer Steel	Open-Hearth Steel	Bessemer Steel	Open-Hearth Steel	Crucible Steel
1849	21,710	....	....	....	....	
1850	39,360	....	....	....	....	
1855	124,000	....	....	....	....	
1860	183,000	....	....	....	....	No Data
1865	318,000	....	....	....	....	
1867	410,000	2,280	....	2,679	....	
1868	445,970	6,450	....	7,589	....	
1869	521,370	8,620	....	10,714	893	
1870	523,000	30,360	....	37,500	1,339	
1875	448,000	260,000	....	335,000	8,080	35,180
1880	441,000	852,000	12,160	1,074,000	110,850	64,660
1885	13,000	959,000	4,280	1,515,000	133,000	57,600
1890	14,000	1,868,000	3,590	3,689,000	513,000	71,200
1895	5,810	1,300,000	700	4,909,000	1,137,000	68,700
1900	695	2,384,000	1,380	6,685,000	3,398,000	100,500
1905	318	3,192,000	183,000	10,941,000	8,971,000	102,200
1906	15	3,391,000	186,000	12,276,000	10,980,000	127,500
1907	925	3,380,000	253,000	11,668,000	11,550,000	131,000
1908	71	1,349,000	572,000	6,117,000	7,837,000	63,600
1909	....	1,767,000	1,257,000	9,331,000	14,494,000	107,400
1910	230	1,884,000	1,751,000	9,413,000	16,505,000	122,300
1912	....	1,100,000	2,105,000	10,328,000	20,780,000	121,500
1913	....	818,000	2,528,000	9,546,000	21,600,000	121,200
1914	....	324,000	1,526,000	6,221,000	17,175,000	89,900
1915	....	327,000	1,775,000	8,287,000	23,679,000	113,800
1916	....	440,000	2,270,000	11,059,000	31,415,000	129,700

(a) In United States—long tons of 2,240 pounds.

christ is credited with making Germany's great industrial development possible.

The well-known "Thomas Slag" which is in demand as a fertilizer on account of its phosphorus content is the by-product of the basic-lined converter.

An idea of what the invention of the Bessemer process meant to railroad development alone may be gained by

Page from "Non-Technical Chats on Iron & Steel" by La Verne W. Spring, showing the drift from Wrought Iron to Bessemer Steel and later from Bessemer Steel to Open Hearth.

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[2] "William Kelly. A True History of the So-Called Bessemer Process" by John N. Boucher. (Published by the author, Greensburg, Pa., 1924), page 89.

Two Converters in Operation and a Third Pouring.

country. However, it was not until 1867 that steel rails began to be rolled commercially. This was at Cambria Iron Company from Bessemer steel blown at the Pennsylvania Steel Co. of Steelton, Pa.

The First Works Laboratory

The Cambria Iron Company had been established in Johnstown, Pa., in 1853. Apparently the men at Cambria were progressive, for they sponsored Kelly's later experiments, at Cambria John Fritz invented the "three-high" rolling mill, and to Cambria also, apparently, must go the credit of having the first iron and steel works chemist. Our well known Robert W. Hunt had spent several years in practical iron making in the rolling mills of John Burnish & Company at Pottsville, Pa. He later took a course in analytical chemistry in the Laboratory of Booth, Garrett and Blair. He entered the employ of the Cambria Iron Co., where on August 1, 1860, he established the first laboratory in America put in as a direct part of an iron or steel plant.[3]

However, in 1861 R. W. Hunt enlisted in the United States Army, served till the end of the Civil War, attaining the rank of captain, and returned to Cambria in 1865. On July 4, 1865, Hunt was made superintendent of the experimental works at Wyandotte. In 1865 Capt. Ward bought the steel works at Wyandotte from the Kelly Process Co. The plant was abandoned in 1869.

In May 1866 Hunt was recalled to Cambria, which intended to build a Bessemer works but this construction was delayed until 1871.

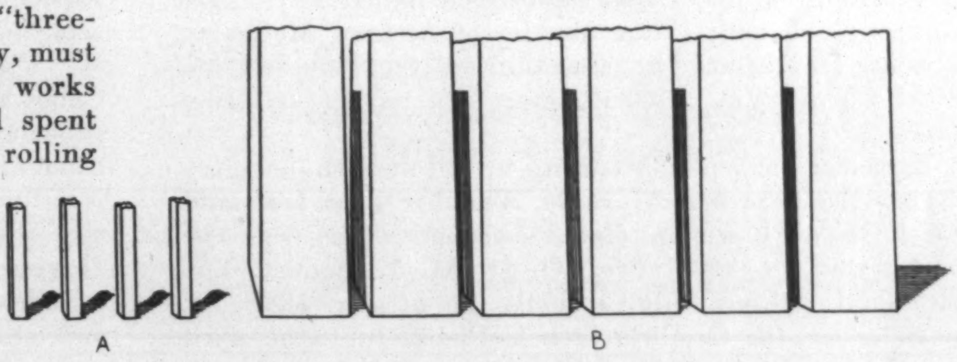
Hunt assisted Geo. Fritz, Chief Engineer of Cambria, in building this plant and took charge on its completion and served until 1873, when he went to the Bessemer plant at Troy, New York. In 1888 he established the Bureau of Inspection and Consultation, now known as R. W. Hunt & Co. Mr. Hunt died in 1923.

Early Bessemer Plants

About the time the Wyandotte plant was being started Alexander L. Holley with John F. Winslow and John A. Griswold, having secured the American rights to Henry's Bessemer's patents, started a 2½ ton converter at Troy, N. Y., in Feb. 1865, under the firm name of Winslow, Griswold and Holley.

It is stated in a history of "Troy and Vicinity" by Arthur James Weise, published in Troy in 1886, that "The first conversion of iron into steel at the Bessemer works was accomplished on Feb. 16, 1865, Crown Point charcoal iron being used. To test the fitness of American iron for conversion into steel, Winslow, Griswold, and Holley wisely experimented with a number of irons manufactured in the United States. As there was no recourse at the time to a chemical analysis of the composition of the iron sent them, many of the brands which were tested were assumed to be unsuited for conversion into steel. 'In the light of our present chemical knowledge of the manufacture,' says a recent writer, 'it is amazing to think of firms sending a few tons to Wyandotte, Troy, or even England to be tried in actual practice, when a few hours of laboratory work would have settled the entire question.' "

The Bessemer Converter is a Prolific Steel Maker.
A. Ingots from 4 pot crucible furnace: each heat 400 lbs. in 4 hours or 100 lbs. an hour. Each heat pours 4 ingots 3x3x36". B. Ingots from one "blow" 15 ton Bessemer Converter; 30,000 lbs. in 20 minutes or 90,000 lbs. an hour. Each heat pours 6 ingots 19x20x62".



CAMBRIA IRON CO. (LATER CAMBRIA STEEL CO.) CHIEF CHEMISTS

	Dates	Educated at
Robert W. Hunt First Chemist, Metallurgist & later Superintendent.	1860-1873	Covington, Ky. Philadelphia, Pa.
Thos. T. Morrell Chief Chemist	Dec. 1867 Aug. 1891	Brown University, 1863
James J. Fronheiser Chief Chemist Supt. Blast Furnace Supt. Metallurgical Dept.	Sept. 1891 Mar. 1894	
Dr. M. E. Rothberg Chief Chemist	Mar. 1894-1900	German Universities
Frank S. Hyde Chief Chemist	1900-1910	Cornell Univ.
Harry S. Hosmer Chief Chemist	1910-1923	Worcester Polytechnic, 1901
Wm. E. Steiner Chief Chemist	1923-Present	Lehigh

SOME OF CAMBRIA'S ASSISTANT CHEMISTS

	Dates	Educated at
James J. Fronheiser Ass't Chemist until 1875. Later, see Chief Chemists.	Until 1875	
Charles L. Taylor Later to Bessemer Steel Works, Homestead, Partner to Carnegie.		Lehigh
E. P. Lee	1880	
John W. Cabot Later Ass't Supt. O. H. & Supt. Bessemer of Bellaire Steel Co.	1879-1888	M.I.T. 1879
M. M. Duncan To Roane Iron Co., Chattanooga then Cleveland Cliffs Co. (1885-1890). Later Gen. Mgr.	1880	Lehigh
Geo. M. Davidson Later Ass't Supt. of Cambria then in 1886 first chemist Engineer of Tests Chicago & Northwestern R. R.	1880	Dartmouth
Webster Norris Later to Commercial Laboratory Pittsburgh. Then 1st Chemist of Chicago Milwaukee & St. Paul R. R. in 1886. Later Consulting Chemist & Eng. for rubber trade.	1881	M.I.T. 1881
Fritz Gleim		German Universities
Clement Le Bouteller Then with Laughlin & Co. as Chief Chemist (1887-1892); then with Taylor Wharton Iron & Steel Co. (1892-1902).	1884-1887	Columbia
Wm. A. Auchinvole		

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[3] A.S.C.E. Memoir prepared by Robert W. Hunt Co., Chicago, Ill.

Although the date is not given it was probably in the early 1870's that: "The laboratory was first in charge of the distinguished chemist, Dr. August Wendel, who by assiduous investigation and careful analyses greatly advanced the manufacture of the best qualities of steel in the United States."

In 1881 J. M. Sherrerd succeeded Dr. August Wendel in charge of the laboratory of the Bessemer works.

In 1866 Winslow, Griswold and Holley at Troy and the Kelly Pneumatic Process Co. at Wyandotte merged, as each owned patents that were essential to successful Bessemer operation. Unfortunately Kelly's name was omitted in any title capacity and thereby America's part in the invention of Bessemer steel has almost been forgotten.

Holley's mechanical inventions in connection with the Bessemer process later put America ahead of the world in Bessemer steel. Holley is credited with designing practically every Bessemer plant built in the United States during the succeeding 10 or 15 years.

From this merger on, progess in Bessemer steel was extremely rapid.

In 1867 the third Bessemer plant in America was built by S. M. Felton at the Pennsylvania Steel Co. according to plans of A. L. Holley. As the rail mill was not yet built, the first steel was rolled into commercial rails at the Cambria Iron Works. In 1868 the Pennsylvania Steel Co. finished its rail mill and Bessemer steel rail making became a well established industry. In 1870 Andrew McCreath (later founder of Andrew McCreath & Sons, Harrisburg, Pa.) established the Pennsylvania Steel Company's chemical laboratory on an enduring basis.

Bessemer steel plants sprang up all over the country during the next several years. The Freedom Iron and Steel Co. at Lewiston (later Burnham, Pa.) was the fourth works (1868), Prof. O. D. Allen, chemist. The Cleveland Rolling Mill Co. built the fifth in 1868. As noted above, the Cambria Iron Co. itself built the sixth Bessemer plant in 1871. Succeeding installations were:

7th Works—Union Iron Co., South Chicago, 1871.
8th Works—North Chicago Rolling Mill Co. (1st blow April 10, 1872) J. H. Cramer, chemist.
9th Works—Joliet Iron & Steel Co., 1873, F. A. Emmerton, chemist.
10th Works—Bethlehem Iron Co., 1873.
11th Works—Edgar Thompson Steel Co., 1873, A. J. Preusse, chemist 1875.
12th Works—Lackawanna Iron & Coal Co., Scranton, Pa., 1875.

The history of the Cambria Iron Co. is interwoven with the names of many men of early Bessemer days who got their first knowledge of Bessemer practice there. Perhaps it might better be said that they had much to do with initiating and getting that practice on an experimental and later on a practical basis.

D. M. Stackhouse of the Cambria plant, now a subsidiary of the Bethlehem Steel Co., has compiled the following interesting sketches of the chief chemists at Cambria and some of the assistants. The earlier chemists at Cambria of course had much to do with the success of the infant Bessemer steel industry.

Cambria Iron Company Chemists

As has been noted above, "Robert Woolston Hunt was Cambria Iron Company's first chemist and probably the first man employed as such by the American iron and steel industry."

"In December, 1867, Mr. Thomas T. Morrell became Chief Chemist and occupied this position until August 31, 1891. He graduated from Brown University in 1863, taught in Bates College prior to going to Johnstown. He took a short post graduate course at Brown University in Metallurgical Chemistry. Mr. Morrell developed new methods for iron and steel laboratory use and was a nationally known authority in this branch of chemistry.

"James J. Fronheiser was assistant to Mr. Morrell until about 1875, when he was put in charge of the Conemaugh blast furnace which made spiegel and ferro manganese. Later Mr. Fronheiser became superintendent of the Cambria blast furnace department and then superintendent of the metallurgical department which included the blast furnaces, Bessemer and open hearth steel works and the blooming mills. Mr. Fronheiser succeeded Mr. Morrell as Chief Chemist in August 1891 and died in 1894 in Germany, of pneumonia, contracted while making a study of by-product coke ovens for the Cambria Plant.

Mr. Morrell's other assistant chemists were Charles L. Taylor, a Lehigh graduate, who later went to the then new Bessemer steel works at Homestead and later retired as one of Carnegie's partners when the United States Steel Corporation was organized; E. P. Lee, John W. Cabot, a graduate of Massachusetts Institute of Technology, 1879 to about 1888, Mr. Cabot later was in the open hearth steel works at Cambria as assistant superintendent, then became superintendent of the Bessemer department of the Bellaire Steel Co., Ohio. George W. Davidson, a Dartmouth graduate, succeeded Mr. Cabot as assistant chemist and later became an assistant superintendent in the Cambria Steel Plant, and then became chemist and engineer of tests for the Chicago and Northwestern Railroad. Webster Norris, a graduate of the Massachusetts Institute of Technology, succeeded Davidson as assistant chemist. Norris later became a member of a firm of consulting chemists in Pittsburgh, and, still later, chemist for one of the large rubber companies. He now is consulting chemist and engineer for the rubber trade. A Mr. Duncan; Fritz Gleim (native of Bavaria and a graduate of German Universities), Clement Le Boutelier (a Columbia graduate), and Wm. A. Auchinvole were later assistant chemists to Mr. Morrell.

Dr. M. E. Rothberg succeeded Mr. Fronheiser as chief chemist in 1894 and left Cambria in 1900 to take charge of the by-product coke ovens at Buffalo, New York.

Following Dr. Rothberg as chief chemists were Frank S. Hyde, 1900 to 1910, a Cornell graduate; then Harry A. Hosmer, 1910 to 1923, a Worcester Polytechnic graduate; then William E. Steiner, a Lehigh graduate, the present chief chemist, from 1923 to present date."

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Fred J. Blaney, for the past 9 years superintendent of the General Alloys Company of Boston and formerly of the Driver Harris Co., Harrison, N. J., has resigned his position to take charge of the Alloy Division of the Ohio Steel Foundry Co., Springfield, Ohio. Mr. Blaney is one of the pioneers in the making of heat resisting alloys, both in the form of rod, wire, sheet and castings, and has come from a family that has been connected with the steel foundries for the past fifty years, and had his training under the supervision of his father who was General Manager of the Benjamin Atha Steel Casting Co., Newark, New Jersey.

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H. J. French, in charge of development work on steel for the International Nickel Company, and a member of the Editorial Advisory Board of Metals & Alloys, has been selected to give the 1933 Campbell Memorial Lecture before the American Society for Steel Treating.

READERS' COMMENT

Editor, Metals & Alloys:

I have read Mr. Pack's interesting article on "Press Casting" in the February issue of your publication. This process has been used for a number of years in Europe for the production of brass castings and is the most successful method yet devised to handle high melting point alloys. The lower melting point alloys such as those of tin, lead, zinc, and aluminum, are, to a very large extent, still produced by the Europeans with standard casting machines, which in many instances, are those of American design.

The contention that a process originally developed for brass will obsolete and supercede all present methods as applied to zinc and aluminum, is neither supported by the short comings of the latter method, nor yet justified by the very limited applications which the press method has had to date.

In modern die cast plants, zinc alloys are cast by the plunger method and the metal is invariably drawn from *below* the surface layer of dross which provides better protection against oxidation than if the metal is ladled out by hand under the press method.

With reference to the alloy, the presence of aluminum in zinc alloys is not the serious problem that faced the industry prior to the intensive research culminated about six years ago. In the old days, the usual standard alloy contained approximately 4% copper, 8% tin, balance zinc. A small amount of aluminum, perhaps 1/4%, was added to facilitate casting. It was not then generally known, that a higher percentage of aluminum in the presence of tin, or minute quantities of cadmium and lead found in the lower grades of spelter was really the cause of rapid deterioration.

The situation now is very different. Zinc is produced commercially with a 99.99 +% purity which a few years back was considered a laboratory curiosity. The objectionable impurities such as those of lead, and cadmium, therefore become negligible. Furthermore, aluminum is added not as a necessary evil, but as a useful element to the extent of 3 to 4% which with a similar amount of copper, forms practically the standard alloy now in use. Tin is no longer used, and to offset the possible effects of traces of undesirable impurities, .02 to .10% magnesium is added as a deoxidizer. It is evident, therefore, that with this knowledge available, even a crude plunger machine can produce reliable die castings, providing it is supplemented with proper gating and venting of dies and "good housekeeping" is maintained to avoid contamination of the alloys.

The last illustration in Mr. Pack's article indicates that even a press casting will show porosity if proper technique is lacking, so after all no equipment can displace the technical casting knowledge necessary to produce the desired results.

Pressure alone, no matter how great, can not eliminate porosity. Those interested in this phase of die castings can find a more complete description as to the causes in the writer's book "Die Casting Practice," McGraw Hill Book Company, 1930.

I am confident that press castings will become a valuable factor in the casting of metals, but will supplement rather than displace die castings. It will find its greatest field of usefulness for brass and magnesium, and only for special applications in zinc and aluminum. Further inroads in the latter field will be governed purely by economic factors, such as the cost of labor and power consumption for a given production compared with those of a modern die casting machine.

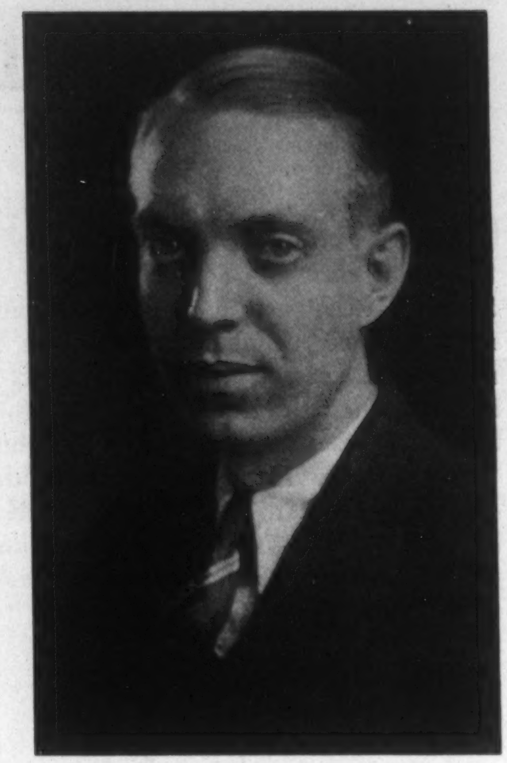
M. Stern
Mgr., Die Cast Division
A.C. Spark Plug Company

March 10, 1932
Flint, Michigan

◆ ◆ ◆

The Copper and Brass Research Association has announced that laboratory work on rapid development of the natural "patina" that gives old copper roofs their attractive green color, has been successful. The development is based on the correction of the former misconception that the patina was a basic carbonate. In fact, it is a basic sulphate. When attention was directed to the production of the basic sulphate and to the conditions under which an adherent coating could be obtained, the natural patina was duplicated.

Development of production methods is under way and it will soon be possible for architects to use spandrels and other decorative pieces made of copper that have, from the erection of the building, the appearance that it has heretofore been necessary to wait for for years while Nature was doing the job.



Edward F. Berry, formerly editor of "Metal Cleaning and Finishing," has resigned that post to become advertising manager of the Udylite Process Company, Detroit, Mich., according to an announcement of L. K. Lindahl, general manager of the Udylite Company.

Mr. Berry is well versed in the field of chemistry, having received the degree of bachelor of science from the University of Pittsburgh in 1926, and then gone with the Aluminum Company of America as research chemist at New Kensington, Pa. He remained there until 1929, when he took over the editorial position mentioned above.

The Udylite Process Company is widely known as licensor of the "Udylite" cadmium plating process for application of corrosion-preventing coatings to metals.

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1932 Annual Convention at Hotel Statler, Detroit, Michigan May 3, 4 and 5

The Board of Directors of the American Foundrymen's Association announces that the next Annual Convention of the Association will be held at the Hotel Statler, Detroit, May 3, 4, and 5, 1932. This announcement supersedes previous announcement that the Annual Convention and Exposition would be held in Philadelphia, May 2 to 6.

The Convention at Detroit will be similar in character to the very successful annual technical and business meeting held at the Edgewater Beach Hotel, Chicago, in 1927, and will be the second time since 1905 that an annual meeting has been held without commercial exhibits.

The A.F.A. also announces that the splendid facilities afforded by the new Convention Hall in Philadelphia have been reserved for the spring of 1933. It is expected that all allied associations and societies which had contemplated meeting jointly with the A.F.A. in Philadelphia will announce that their meetings will be held during the days of the A.F.A. Convention in Detroit.

The Program and Papers Committee of the A.F.A. has been unusually fortunate this year in having a wealth of good papers from which to select in organizing the A.F.A. program. These sessions, together with those of other organizations meeting jointly with the A.F.A. at that time, will provide such a concentrated meeting as will enable foundry managers, engineers, metallurgists and shop men to get the most value for a minimum of expense and time.

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B. F. McIntyre is now connected with the Chicago office of The Lincoln Electric Company, manufacturers of "Linc-Weld" motors and "Stable-Arc" welders, in charge of motor sales for the Chicago district. McIntyre was formerly with the Industrial Controller Company of Milwaukee. His electrical engineering education was obtained in Toronto, Canada.

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Ralph Fink has joined the laboratory force of the Ferro Enamel Corporation, Cleveland, Ohio. Mr. Fink is a graduate of Ohio State University, having received an A.B. in Chemistry there. He also studied in Cleveland College, majoring in Metallurgy of Iron and Steel. He was formerly with the American Stove Company of Cleveland.

MOLYBDENUM—

By J. L. GREGG* and

ONLY a few years ago molybdenum was classed as a "rare" element, but the discovery of large domestic deposits made this a misnomer long before the metal achieved general commercial use. Within a very brief period it became one of the better known alloying elements, and its use was officially recognized in 1926 when the Society of Automotive Engineers added the "4100" series (chromium-molybdenum) to its lists of common alloy steels. After molybdenum had been successfully used in forged and rolled steels it was found that in combination with either nickel, chromium or manganese it was decidedly useful in cast steels, and consequently it has been used in a large tonnage of steel castings during the past few years. Serious attempts to use molybdenum in cast iron, outside of rolls, have been made only within the past two or three years, but the properties of molybdenum cast iron are such as already to have warranted its commercial application for a number of important services. It appears that its use in cast iron will be greatly extended.

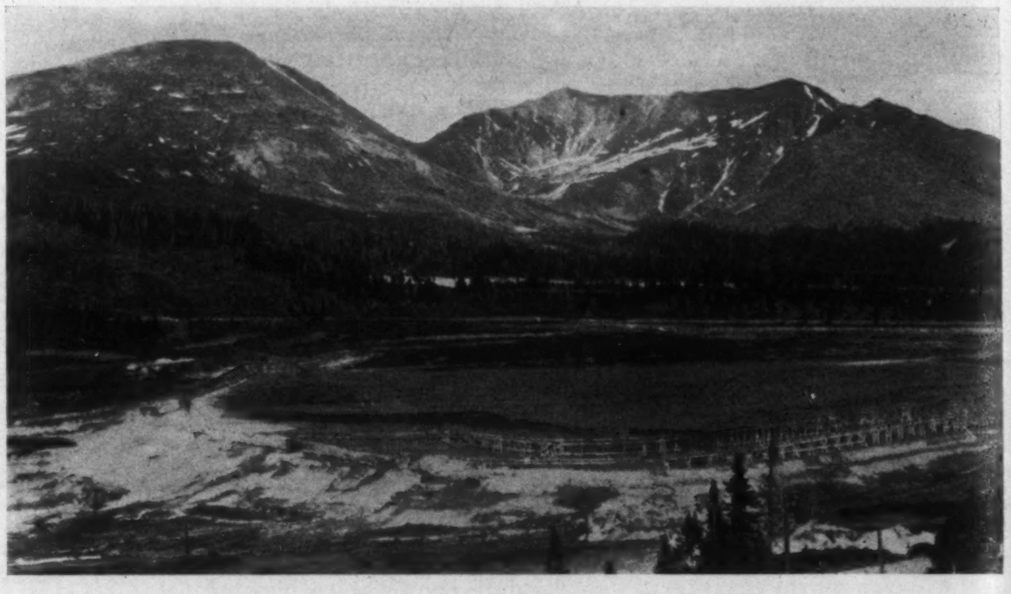
An effect of molybdenum in steel and cast iron which has only recently been given the recognition and emphasis it deserves, is its specific effect in increasing resistance to creep at high temperatures. Its effect in this instance is truly phenomenal, for the action of a few tenths of a percent of molybdenum in increasing the resistance of steel to creep at temperatures as high at 1000° F. is much greater than the effect of several percent of other alloying elements frequently used. Most of the commercial nitriding steels contain a small quantity of molybdenum, whose chief function is to prevent temper brittleness that might arise from the nitriding operation.

A large tonnage of molybdenum is used in tool and die steel, principally in die steels used in hot workings or those subjected to particularly severe service. A larger tonnage of molybdenum, however, is probably used in rolls, both steel and iron.

AUTOMOTIVE ENGINEERING STEELS

Molybdenum, early in its commercial history, came into prominence as an ingredient of rather highly alloyed steel, such as the 3% nickel, 0.9% chromium, 0.5% molybdenum steel used for Liberty Motor crankshafts. In such steels the effects of molybdenum in allowing high tempering temperatures for a given hardness, in producing depth-hardening on quenching large pieces, and in avoiding temper brittleness were quite soon established. The general acceptance of molybdenum as an alloying element, however, resulted from the successful use of chromium-molybdenum steels (the 4100 series) in such vital parts of the automobile as steering knuckles, steering arms and rear axle drive shafts. It is generally recognized that the mechanical properties of the 4100 series of steels are comparable to those of more highly alloyed materials, and that the use of these steels in place of the latter materials frequently results in considerable saving without sacrifice in the quality of the product.

Shop experience with molybdenum steels has shown that, with the same Brinell hardness, molybdenum steels are easier to machine than other alloy steels, although laboratory tests by Digges[3] have indicated that for shallow cuts this does not hold for all hardnesses. It is always difficult to express machinability in acceptable quantitative terms, but a machine tool builder would appear to be a competent judge of the problem. Hence the following statement by D. L. Gurney of the Warner-Swasey Co. made at the Nov. 1931 Conference on Metals and Alloys at the Case School of Applied Science is instructive.



Out in the Molybdenum Country—View of a Tailings Pond.

> "S.A.E. 4140 steels find their application for parts that are to be machined after heat treatment. Chromium-molybdenum steels are unique in that they possess exceptional machining qualities at higher Brinell hardness values than any of the usual alloys. It is not uncommon to machine this grade of steel at Brinell values ranging from 321 to 418.
>
> "This steel is used for intricate parts requiring high hardness and high physical properties that would be difficult to obtain without undue distortion using some of the other steels, heat treated subsequent to machining."

Like machinability, wear resistance is a property that is difficult to determine in the laboratory. It is claimed that molybdenum increases the wear resistance, and the good service of molybdenum steels in applications where failure by wear is a factor backs up this claim. Such applications include the use of the steel in shovels, dipper teeth and the renewable wearing parts of various types of rock-crushing, grinding and handling machinery in the cement, mining and allied industries. The function of molybdenum in these steels is to produce toughness combined with high hardness, such a combination reducing the danger of breakage in service of this type.

After chromium-molybdenum steels had been established in automotive use their particular fitness for aircraft tubing made them standard for this application. Here their advantage lies chiefly in their ease of welding and the good properties of the weld and the adjacent section, on mere air-cooling, in the dimensions used in such tubing. Chromium-molybdenum steels, according to

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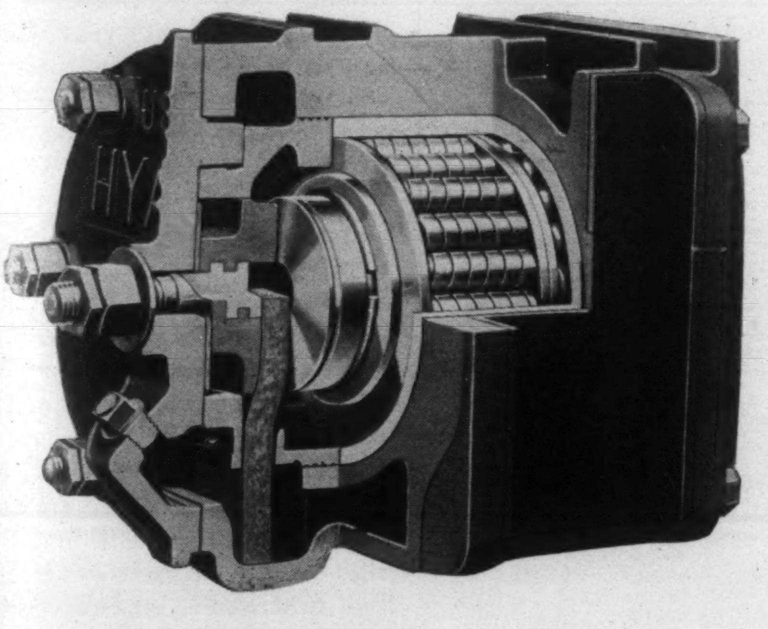
*Battelle Memorial Institute, Columbus, Ohio.

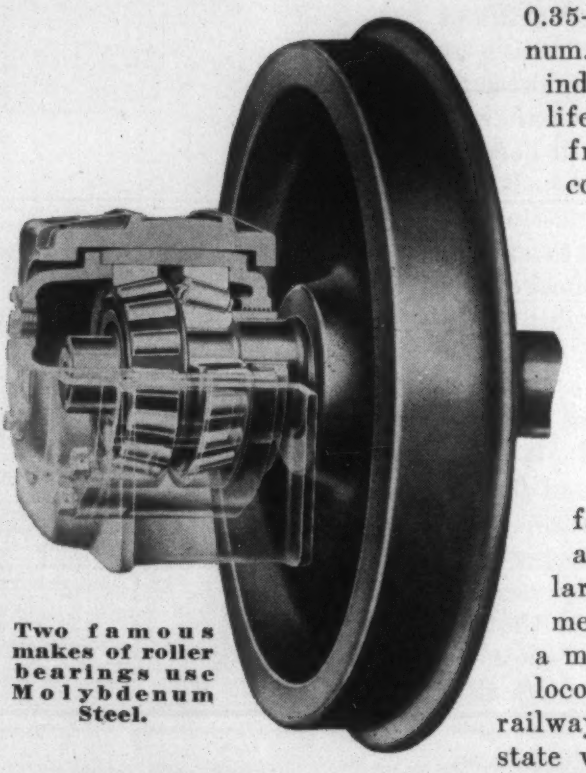
Today and Tomorrow

H. W. GILLETT*

a private communication, have recently been used by the Winchester Repeating Arms Co. for the barrels of shotguns, which are heat treated after forging to give a tensile strength of from 115,000 to 125,000 lbs./in.^{2}, and an elongation of from 25-30%. The frame and a number of the working parts are also made from a chromium-molybdenum steel, the frame being treated to a tensile strength of approximately 200,000 lbs./in.^{2} The free machining property of this type of steel is particularly advantageous on account of the high degree of accuracy necessary in such parts.

Although the chromium-molybdenum type of steel is more generally used for structural parts than any other molybdenum steel, particularly in this country, the extensive use of nickel-chromium-molybdenum steels in the "Bluebird"[4] shows that this latter type of steel is relied upon where a steel of very high quality is essential. In a recent book by Macrae[6] on the auto-frettage process of cylinder and gun construction used at Woolwich Arsenal most of the steels discussed are of the nickel-chromium-molybdenum type.





Two famous makes of roller bearings use Molybdenum Steel.

According to Davis,[2] American manufacturers of ball bearings use a steel containing 0.95-1.10% carbon, 1.20-1.50% chromium and 0.15-0.30% molybdenum. For carburized cases a nickel-molybdenum steel is used containing 1.5-2.0% nickel and 0.20-0.30% molybdenum, which carburizes uniformly, is free from soft spots and has a tough core. This type of steel, S.A.E. 4615, is extensively used for carburizing in this country, and, according to Rowe,[10] it is also finding use in England. It is particularly attractive because good properties in both case and core can be obtained by a single quench, as mentioned in the 1931 National Metals Handbook. Manganese-molybdenum steels are likewise used for carburizing, although but meager data are to be found in the literature regarding the properties of these steels when carburized.

A steel producer who specializes in alloy steels has done considerable experimental work on manganese-molybdenum steels, but the results of this study have not yet been reported. Lt. Linke, of the U. S. Navy, has recently studied the properties of a forged manganese-molybdenum steel, and his tests on a steel containing 0.27% carbon, 2.67% manganese and 0.44% molybdenum showed that this steel had markedly good properties after quenching and tempering.

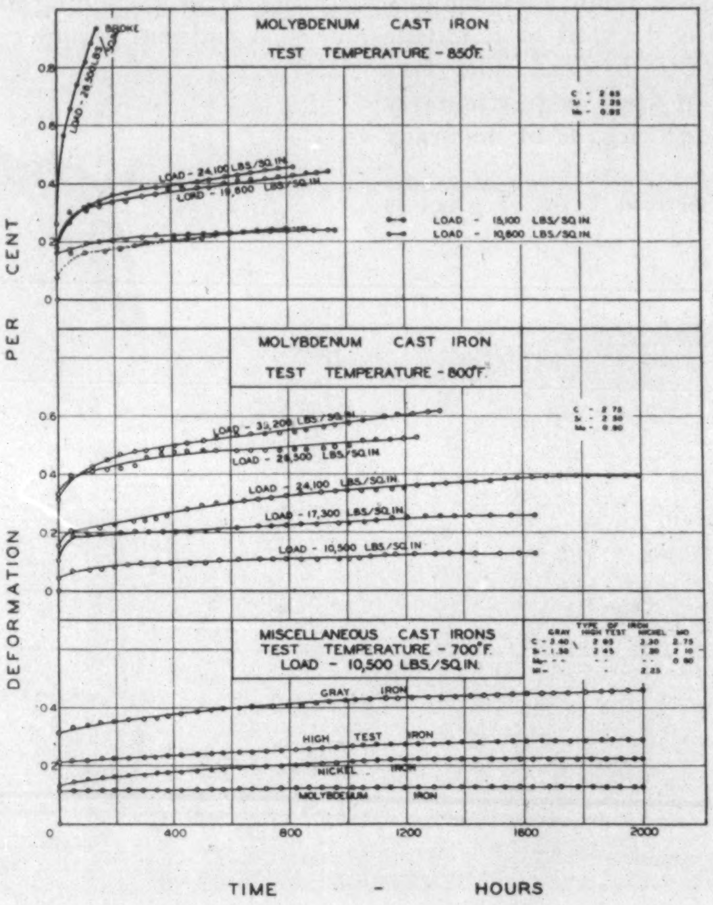
A railroad is now conducting tests on a rail containing 0.60-0.75% carbon, 1.45-1.75% manganese and 0.35-0.45% molybdenum. The tests have indicated that a long life can be expected from a rail of this composition, for a 90 lb. rail in severe service since 1927 indicates that its life will be a little over 4 years while the usual type of plain carbon open-hearth rail fails after a year and a half of similar service. Pohl[9] mentioned the use of a molybdenum steel in locomotive and street railway tires, but did not state whether or not the steel contained other alloying elements. The superior wear resistance of the molybdenum steel is said to increase the life of both the tires and the track.

The recent work of Burns[1] reported at the last meeting of the British Iron and Steel Institute shows that the manganese-molybdenum steels of the structural type have considerable merit. It was found that additions of up to 0.3% molybdenum to steels containing 1.3% manganese

Table I. Properties of Molybdenum Steels. (Burns[1])
Treated to Give a Brinell Hardness of 250

No.	KEH	KEJ	KEK	DKH
Composition, %	C 0.29 Mn 1.32 Mo Nil	C 0.35 Mn 1.24 Mo 0.22	C 0.32 Mn 1.22 Mo 0.32	C 0.33 Ni 2.70 Cr 0.58 Mo 0.57
Treatment:				
Hardening Temperature °F.	1560	1560	1560	1650
Hardening Medium	Water	Water	Water	Oil
Temper °F.	1020	1130	1150	1240
Hrs. Tempered	2	2	2	3
Tensile Tests:				
Elastic Limit, lbs./in.^{2}	67,200	76,200	91,800	98,600
Yield Point, lbs./in.^{2}	83,300	96,800	100,400	105,100
Tensile Strength, lbs./in.^{2}	107,700	115,800	116,500	121,200
Elongation, %	24	24	24	23
Reduction of Area, %	67	67	67	64
Torsion Test (0.4" diam. parallel 5")				
Elastic Limit, lbs./in.^{2}	47,000	47,000	56,000	62,700
Max. Shear, lbs./in.^{2}	101,900	103,900	101,900	108,900
Degrees of twist per in. length	400	310	355	400
Brinell Hardness	249	252	256	254
Izod Impact, Ft.-lb.	62	75	82	61
Sankey Bend Test:				
ft.-lb.	2962	3516	2733	3560
Number of Bends	26	31	24	27

had little effect on the properties of the unheat-treated steel, but that it had a marked beneficial effect on the steel in the hardened and tempered condition. In steels containing 2.3% manganese the addition of 0.5% molybdenum produced a steel which when properly hardened gave mechanical properties approximating those of a nickel-chromium-molybdenum oil hardened steel. The properties, as determined by Burns, of one manganese, two manganese-molybdenum steels, and a nickel-chromium-molybdenum steel all heat treated to a Brinell hardness of 250 are shown in Table I. Sometime ago it was found that the addition of several tenths of a percent of molybdenum to nickel-chromium steels prevented the appearance of temper brittleness, but that increased manganese favored the propensity to brittleness. Before using a hardened and tempered steel containing as much as 2% manganese it is well to inquire whether or not such a steel will become brittle on tempering. Burns found that the addition of 0.5% molybdenum to a 2.3% manganese steel almost eliminated the susceptibility to temper brittleness. He also found that a steel containing 0.32% carbon, 1.34% manganese and 0.21% molybdenum had good depth-hardening properties.



Before molybdenum became to be generally used in steels it was known that it had a marked tendency to decrease the "mass effect," and in fact its use in a number of steels is predicated on this behavior. Recent work of Kallen and Schrader[5] has afforded additional confirmation, and has definitely shown that ¼% molybdenum in a steel containing 1% chromium is of decided benefit in producing hardening throughout the entire section of bars up to 4 inches in diameter.

The use of a steel containing 0.25-0.35% molybdenum and 0.10% vanadium for centrifugally cast guns has been continued at Watertown Arsenal, and it is claimed that this type of steel can successfully replace more highly alloyed steel forgings, for such use as in axles.

The utility of molybdenum in combination with other alloying elements in steel is so thoroughly recognized that few present-day alloy steel patents neglect to point out that the good qualities claimed for the composition under discussion are preserved in the presence of, and enhanced by, the addition of a little molybdenum.

Thus molybdenum, in the combinations mentioned above and in others not yet so thoroughly studied, seems to be compatible with all the other common alloying elements in steel and, broadly speaking to confer upon such alloy steels some property—often a most desirable one, that either could not be obtained at all without it, or would require the use of a much larger quantity of some other element. This fact has apparently intrigued most investigators, for, save a very few of the early students of the effects of molybdenum in steel, such as Swinden, later workers have badly neglected the consideration of steels containing only molybdenum as an alloying element. Indeed the so-called molybdenum steels of commerce usually contain at least one alloying element in addition to molybdenum.

Recently, however, a low carbon steel containing only molybdenum as an alloying element has found use abroad for boiler construction. This will be discussed later in this article. Such a steel has remarkable strength at high temperatures, and can be fabricated as readily as the unalloyed boiler steel.

STEELS FOR HIGH-TEMPERATURE SERVICE

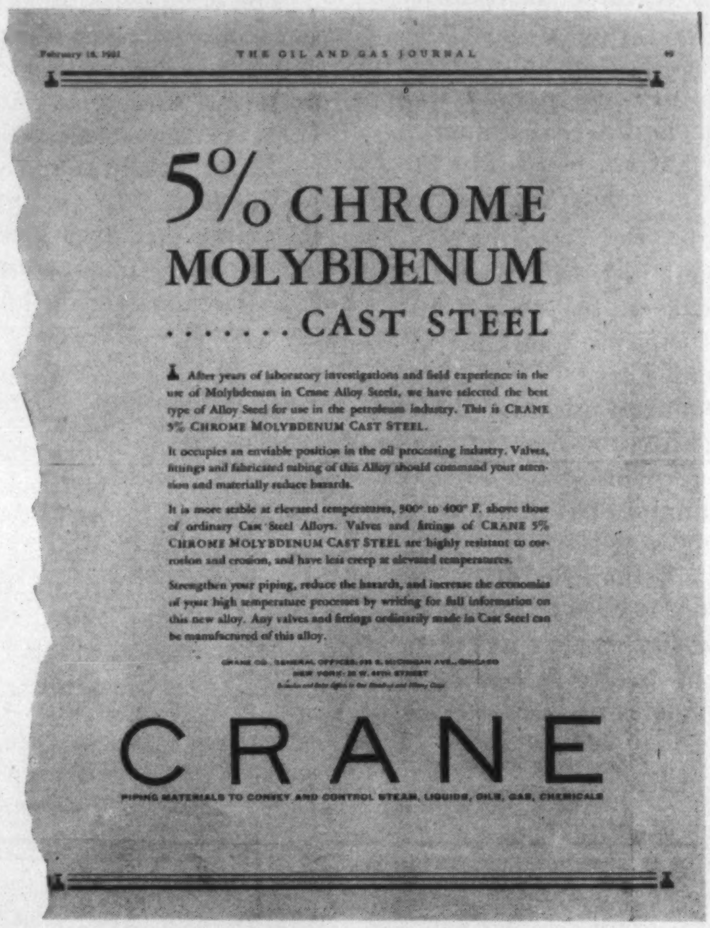
Within the past few years there has arisen a demand for steels having high resistance to creep at elevated temperatures, and as the properties of more and more steels were studied at elevated temperatures it became apparent that molybdenum has a specific effect on a broad range of steels in raising the resistance to creep. The past year has shown a markedly increased realization of this property of molybdenum as is proved by references to its utility scattered all through the 1931 A.S.T.M.-A.S.M.E. Symposium on Effect of Temperature on Metals.[23] Whether in a carbon, a nickel, a chromium or a nickel-chromium steel, molybdenum generally boosts a steel that would otherwise be classed as mediocre, up to an outstanding position for high-temperature service. The best choice and amount of other elements to accompany it are not yet fully determined, and there is something of a conflict between tungsten and molybdenum (which of course, from their position in the Periodic Table have somewhat similar properties) with molybdenum in the lead, according to Bailey, Dickenson, Inglis and Pearson.[11] These English authorities represent groups not only highly experienced in the development of alloys for high-temperature use, but also in engineering design of high-pressure steam power plant equipment and apparatus for use in fixation of atmospheric nitrogen.

Such user testimony seems significant. The statement by Tapsell[24] of the National Physical Laboratory in his recent book on "Creep in Metals" that "there appears to be some agreement that considerable improvement in resistance to creep may be effected by the addition to carbon steels of molybdenum and of nickel, chromium, and molybdenum in combination" is also significant.

The work of Prömper and Pohl[19] in 1928 quite definitely proved that at temperatures up to 950° F. low-carbon steels containing several tenths of a percent of molybdenum had decidedly higher so-called creep limits than unalloyed steels, and that these low-carbon molybdenum steels could be readily fabricated, even when fabrication involved welding. From their work it was apparent that a steel containing 0.15% carbon and 0.39% molybdenum will permanently withstand a decidedly greater load at 950° F. than a carbon steel containing twice as much carbon. These findings have been confirmed by other investigators, and a steel of this type is being used abroad,[20] e.g., in Loeffler boilers. This type is being studied in the United States with a view to its use in pressure vessels at high temperature, though it has not so far been widely used in this country. The steel used in Loeffler boilers, containing 0.25% carbon, and 0.6% molybdenum, was found by the Vitkovice Works to have the following properties:

Tensile Strength at 70° F. 106,300 lbs./in.^{2}
Tensile Strength at 1020° F. 63,100 lbs./in.^{2}
Yield Strength at 70° F. 72,100 lbs./in.^{2}
Yield Strength at 1020° F. 58,300 lbs./in.^{2}
Elongation in 4 in. at 70° F. 14%
Elongation in 4 in. at 1020° F. 18%

A number of important references to the high-temperature properties of various steels containing molybdenum are given in the selected bibliography following this article. An interesting proof of the suitability of molybdenum steels for high-temperature service is offered by Jungbluth,[14] who describes an experiment at the Krupp laboratory where bars of various alloy steels were loaded, held at a temperature of 930° F., and their strength at this temperature judged by the permanent set of the samples. The steels used, which included nickel, and nickel-chromium steels could be readily divided into two classes; i.e., those containing molybdenum which had the higher resistance to flow at the temperature of test and those not containing molybdenum which were inferior.



THE OIL AND GAS JOURNAL

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CRANE

PIPING MATERIALS TO CONVEY AND CONTROL STEAM, LIQUIDS, OILS, GAS, CHEMICALS

So much work has been reported on the high-temperature properties of molybdenum steel within the past two years that it is impossible to review even the outstanding work in this brief summary. However, almost without exception the published work indicates that molybdenum, either in wrought or cast steel, has a remarkable effect in increasing the strength of the steel at medium high temperatures and that the only rivals of molybdenum in this field are tungsten and vanadium. As Dr. J. A. Mathews[18] points out "A single addition of chromium to steel does not seem to give high creep strength. H. D. Newell in a recent paper before the American Petroleum Institute gives creep values for 4 to 6% chromium steel even lower than for plain carbon tube steel at 900° F., but approximately equal to plain carbon steel at 1000° F. Resistance to creep in this steel can be very materially improved by small additions of either tungsten or molybdenum."

Cast valves and fittings of the 5% chromium steel plus molybdenum are, after long development and much practical experience, now being advocated by one of the largest producers as "the best type of alloy steel for use in the petroleum industry" for valves and fittings.

While much less information is available on the effect in cast iron, a very similar behavior to that of molybdenum steel is met when molybdenum is added to cast iron, for Lorig and Dahle[42] found the resistance to creep of molybdenum cast iron at 700° F. to be appreciably greater than that of an unalloyed high-test iron or of a nickel-iron. Since that article was published they have continued the creep-testing of molybdenum cast iron at still higher temperatures and some of the results obtained by these workers at the Battelle Memorial Institute are shown in the diagram on the facing page.

SPECIAL STEELS

A development of rather recent times is the recrudescence of molybdenum in high-speed steel, either tungsten-free or with a minor tungsten content. Capt. Ritchie[33] of Watertown Arsenal has shown that a borax coating will prevent the decarburized bark (often ascribed to volatilization of molybdenum) previously considered to be a difficulty not readily surmounted. The tool performance is at least sufficiently creditable to give such steel a definite place in the sun, on the basis of work per dollar of steel cost (since the steels contain only 8 or 9% Mo), quite irrespective of the urge to prove the feasibility of using a domestic alloying element in place of an imported one, which actuated the War Department in its work. The Ludlum Steel Co. made a regular sized electric furnace heat of this steel during the past year, and according to information from one of the officers of the company, no difficulty was encountered in making the steel. The steel produced was structurally sound and from the recovery standpoint equally as good, if not better, than regular "18-4-1."

Some of the new high-speed steels now contain 1 or 2% molybdenum, which, according to Oertel and Grützner[31] increases cutting life, increases hardness and promotes solution of the carbides. Molybdenum hack-saw blades are on the market, and we may look for a slow but steady development of various molybdenum high-speed steels for some types of tools.

Molybdenum is quite generally used in complex tool and die steels, but little has been published on molybde-

Table II. Properties of Heat-Treated Molybdenum Steel Castings of the Approximate Compositions:

Grade A 0.35% C, 1.60% Mn, 0.35% Mo
Grade B 0.35% C, 1.50% Mn, 0.65% Cr, 0.35% Mo
Grade C 0.35% C, 1.35% Mn, 1.15% Ni, 0.60% Cr, 0.35% Mo

Grade	Yield Point lbs./in.^{2}	Tensile Strength lbs./in.^{2}	Elongation in 2 in. %	Reduction of Area %	Brinell Hardness	Charpy ft.-lb.
	Air Cooled from 1650° F., and Tempered at 1250° F.					
A	68,000	96,400	25.4	57.6	211	23.0
B	81,000	103,750	23.1	55.8	238	19.8
C	88,200	110,400	21.9	49.5	258	21.4
	Oil Quenched from 1575° F., and Tempered at 1250° F.					
A	76,400	102,300	25.8	58.1	228	29.0
B	93,500	113,500	23.1	56.1	257	26.2
C	90,700	120,900	21.3	44.1	262	22.0

num for such uses. The same is true of rolls, although Hruska[27] has given a good discussion of the composition of both iron and steel rolls. In a recent trade publication the United Engineering and Foundry Co. advertised a chilled iron roll containing from 0.20-0.40% molybdenum, which is applicable to the usual duties of chilled rolls, but is tougher, finer grained and has a better surface. The time required for turning the roll is not increased by the addition of molybdenum.

Automotive valve steels frequently contain molybdenum in addition to other elements, and it is quite difficult to determine whether or not molybdenum serves a useful purpose in such steels. In fact some makers of valves do not believe that molybdenum is of value. A recently conducted laboratory investigation by Losana,[29] however, in which the rate of scale formation of various steels when heated in air in products of combustion was studied, indicated that molybdenum had a real influence in decreasing the rate of scale formation. Certainly the effects of molybdenum in steels for exhaust valves are worthy of further investigation.

We probably hear a lot more about nitriding than the present use of nitrided parts might seem to justify, but this is a result of the belief of the engineer and metallurgist that nitriding is going to find a large future use. Useful nitrided cases have been produced on only a limited number of alloy steels, and the alloying element generally considered most effective in promoting the formation of a satisfactory case is aluminum. In addition to aluminum most commercial nitriding steels contain both chromium and molybdenum. The outstanding effect of molybdenum in this type of steel is in decreasing brittleness in both case and core. This, like its effect on temper brittleness in other steels, is again pretty much of a specific effect, difficult to get without, and easy to get with, molybdenum.

Another group of alloys deservedly in the public eye is the "18-8" type, including those of higher nickel or chromium content. To these austenitic corrosion-resistant alloys molybdenum in quantities up to 2 or 4% has recently been added for quite definite purposes. As mentioned by Aborn and Bain,[26] molybdenum increases the resistance of the alloy to sulphurous acid under pressure, and is therefore of value in paper mill machinery. Molybdenum also tends to decrease the susceptibility of this group of alloys to intergranular deterioration at elevated temperatures. Although other elements have been used as a preventive or palliative against intercrystalline embrittlement there appears to be an increasing tendency to add both molybdenum and copper. While the mechanism of this type of failure is not yet entirely satisfactorily explained, and one cannot yet safely predict what the final answer to the problem will be, it has been pretty clearly shown that 1% or more molybdenum in conjunction with a small amount of copper has a definite effect in lessening the susceptibility to this type of failure.

Another important corrosion-resistant alloy recently used is a nickel-base alloy containing an appreciable quantity of molybdenum, which is resistant to hydrochloric and sulphuric acid.[49] Again, this is so far a minor use of molybdenum from the tonnage point of view, but the corrosion troubles in some chemical industries are far from minor and anything tending to relieve them is quite worth recording.

Molybdenum has thus ramified its use into many of the complex and special steels for very diverse purposes. The fields of application to be mentioned below have not yet extended quite so widely, but even here the boundaries are rapidly spreading out.

STEEL CASTINGS

After molybdenum had been found to be a desirable alloying element for many types of wrought steels it was in order to attempt to use it in cast steel, and it is now being used in castings by many foundries. As in the case of wrought steel, its advantages are most evident when used in conjunction with other alloying elements. The effect of molybdenum in increasing the response to heat treatment and in promoting depth-hardening makes it an ideal alloying element for castings that are to be heat treated.

The compositions of molybdenum cast steels vary widely. Kinnear[36] has recently reported that desirable properties can be produced in an alloy containing approximately 1% copper and 0.35% molybdenum by a simple heat treatment. An Ohio foundry produces a cast steel containing 0.25-0.35% carbon, 0.90-1.25% manganese and 0.25-0.35% molybdenum, which is made either in the open hearth or in the electric furnace. After quenching in water from 1600° F. and tempering at 1250° F. this steel has the following properties:

Tensile Strength	100,000 lbs./in.^{2}
Yield Strength	75,000 lbs./in.^{2}
Reduction of Area	60.5%
Elongation in 2 in.	26.0%
Endurance Limit	53,500 lbs./in.^{2}

A Midwestern foundry produces steel castings containing 0.50% carbon, 1.50% chromium and 0.60% molybdenum, which are used where a high resistance to wear is important, as in dipper teeth. As normalized this steel will have a Brinell hardness of 400 in a section 2 or 3 inches thick. Another Midwestern foundry has experimented with the following three grades of molybdenum steel containing about 0.35% carbon:

Grade A
1.60% Mn and 0.35% Mo
Grade B
1.50% Mn, 0.65% Cr and 0.35% Mo
Grade C
1.15% Ni, 0.60% Cr, 1.35% Mn, 0.35% Mo

Excellent physical properties were produced by heat treatment in all three types of steel, and grade C was adopted as one of their standard steels. This steel will develop hardness values on air cooling that can be produced in some of the other steels only by an oil or water quench, and this property makes it particularly valuable for use in intricate castings that are not well suited for quenching in a liquid. After normalizing and tempering, no difficulty was experienced in meeting the following minimum values:

Tensile Strength	115,000 lbs./in.^{2}
Yield Point	85,000 lbs./in.^{2}
Reduction of Area	40%
Elongation in 2 in.	20%

The properties of the three grades of steel mentioned above when normalized or quenched in oil and then tempered at 1250° F. are listed in Table II. Decidedly greater strengths with only slightly lowered ductility values can be produced in these steels by tempering at a lower temperature.

CAST IRON

Most of the applications of molybdenum mentioned above have followed quite naturally as the specific effects of the element in steel became better understood, and have been gradual developments over quite a period of years. But from the point of view of real metallurgical "news," perhaps the outstanding developments have not been in steel, but in cast iron. Although molybdenum has been used in chilled iron castings, such as rolls, for some time, its use in gray iron is a quite recent development. A recently issued booklet[39] has described the properties of a number of molybdenum irons and has called attention to the already extensive use of molybdenum cast iron in the automotive industry. In automotive irons molybdenum is, so far, used either alone or in conjunction with nickel and less frequently with chromium or both nickel and chromium. The molybdenum content of such irons varies from 0.35 to 0.75%, depending on the properties desired in the iron and the base composition. Not infrequently equal amounts of nickel and molybdenum are added.

The effect of molybdenum on the properties of the iron depends, of course, on the type of iron and on the foundry technique, but a survey of the published data on the properties of iron containing molybdenum indicates that with the addition of 0.5% of molybdenum to gray iron the transverse strength is increased from 15 to 25% and the tensile strength from 20 to 30%.

An interesting application of molybdenum cast iron has been for brake drums. The advent of free wheeling and the increasingly high speeds at which modern cars can be comfortably driven require that they be also capable of stopping quickly. The cast iron drum followed as a natural sequence and on one of the most powerful cars molybdenum iron has been very successful. Relatively little service experience has been had on any car as yet, and it is not yet clear whether molybdenum cast iron will be found essential and become standard on lighter cars. But if the high temperature properties of the iron are an indication of the sort of thing that is needed when we jam on the brake pedal, and since the braking surface probably does get rather hot, the chances would seem to favor this important new field being pretty well commanded by molybdenum.

The remarkable uniformity of strength obtained in large sections of molybdenum cast iron is evident from Mackenzie's[43] tests of the properties of bars cut from a 10 inch diameter section of an iron casting containing 0.90% molybdenum, with 3.05% carbon, 1.22% silicon, 1.16% manganese, 0.30% phosphorous, 0.09% sulphur and traces of nickel, chromium and vanadium. A separately cast bar gave 13,000 lbs. load, 0.21 in. deflection, while six 2.20 in. diameter bars (24 in. span) cut from the 10 in. diameter casting gave from 12,000 to 13,000 lbs. load and 0.29 to 0.33 in. deflection. Tensile tests of bars from near the edge and from the center gave, respectively 39,000 and 38,600 lbs./in.^{2} A diagram showing the details of each test is given by Mackenzie[43] and by Lorig and Dahle.[42]

The many recent investigations of the properties of molybdenum cast iron have proved that a small quantity of molybdenum notably strengthens the iron without rendering machining particularly difficult. The results of Musatti and Calbiani[46] indicate that molybdenum very appreciably increases the wear resistance of cast iron, and service results appear to support this conclusion. From a careful review of the published work on molybdenum cast iron and from the evidence of the work now in progress at Battelle Memorial Institute we believe that molybdenum is definitely in the picture, and that its use in cast iron is quite as logical as its use in steel. As in the case of steel, it is probable that molybdenum will find its greatest use in conjunction with other alloying elements, and quite likely will find itself in about the same company. While plain molybdenum iron is of obvious value, and work on molybdenum properly starts with such an iron, it is likely that, as a competitor of nickel and chromium in cast iron, it will turn out to be a very friendly one, since metallurgical reasoning indicates that a suitable combination of these three elements may produce better results than one or two of them. This is more of the order of prophecy than history, but it is in line with recent comments by Stanley.[49] At any rate molybdenum is very definitely in the alloy cast iron picture, and we believe, there to stay.

Although molybdenum is not generally used in malleable cast iron Jennings and Henderson[40] have shown that, while it decreases the rate of graphitization, it does not prevent it until a content of 5% is reached, and they hint that improved toughness and strength may be imparted by smaller amounts. Boegehold's patent (U. S. No. 1,707,753, 1929) also indicates that molybdenum may be a desirable alloying element in malleable iron. Some unpublished work in Germany has shown that molybdenum in white-heart malleable iron increases the strength of the iron without interfering with the annealing process. Future findings with respect to the effects of molybdenum in malleable iron will be awaited with interest.

STATUS OF MOLYBDENUM

The present status of an alloying element should not be evaluated on the tonnage used during the past two "lean" years, but rather on the extent to which the element entered a broader field of usefulness. From this basis the position of molybdenum has quite definitely improved for it has achieved an increasing importance as an alloying element in all classes of ferrous products.

One might comment that molybdenum has a decidedly good reputation as an alloying element in the public eye, for when a manufacturer starts to use molybdenum steel in his products, he is very likely indeed to advertise that fact instead of merely adopting some fancy brand name that gives no clue as to what makes the steel a good one for his purpose. This is doubtless because of the very good record of molybdenum steels in licking tough jobs and their suitability for severe service.

There has been no diminution in the amount of technical information progressively available in scientific and engineering publications on the effect of molybdenum as an alloying element. Some of the important recent articles are listed in the appended selected bibliography, but no detailed excerpts need be given here. The 1929 and 1930 articles have been abstracted in pamphlets issued by one of the producers of the metal,[50, 51] and the whole subject, including the 1931 literature, is summarized in the monograph on Iron-Molybdenum which has been prepared at Battelle Memorial Institute in coöperation with the Alloys of Iron Research Committee, and which will soon go to the printer.

Considering the relatively short time that molybdenum has been in commercial use, it has had an astonishingly rapid rise to a very important position in the group of useful alloying elements. Its position there seems secure, not only for the present but for the future. The more the ferrous industries know about it, the less easily can they do without it.

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EDITORIAL COMMENT

(Continued from Page 83)

shooting" can be justified even when assayed over comparatively short time intervals. The job of providing business insurance is a long-time job which cannot be assayed over intervals of less than five years, and sometimes must be assayed over ten year intervals. It is not so much the purpose of a research laboratory to enable the stockholders to *get* more money as it is to enable them to *earn* more money. New fields must be found for existing products; better and cheaper products and processes must be found for old fields; new products must be found for new fields. This is another way of saying that there is more competition between industries than there is between individuals in the same industry. An industrial research laboratory is only a reasonable insurance against loss in this sort of competition.

A research organization can not be bought off the shelf like groceries—it must be a product of years of growth and evolution inside the company organization. There is the same difference between a true research organization and a group of scientists newly gathered together that there is between an army and a mob, or that there is between a sales force and a lot of pack peddlers. It is as short sighted to wipe out a research organization as it would be to discharge the sales force. Neither can be duplicated as an efficient working unit on short notice.

The serious attention of business management is directed to the following proposal for a rational order of retrenchment in hard times:

> First—reduce factory expenses in all reasonable ways;
> Second—reduce and, if necessary, omit dividends on the common stock;
> Third—reduce dividends on the preferred stock;
> Fourth—omit dividends on the preferred stock;
> Fifth—reduce the laboratory appropriation;
> Sixth—wipe out the laboratory and turn the company over to the bondholders.

When the laboratory goes, the insurance on the future of the business goes. Next to meeting the interest on the bonds, and the maintenance of an adequate sales force, the first obligation of management to the stockholders is the maintenance of an adequate research organization. It is an obligation greater than that of paying preferred dividends, for it has to do with insurance against ruinous inter-industry competition in the future.

The stock of a company which has seriously cut into its research and development program in order to pay preferred dividends is a poor gamble. If the laboratory is going full blast the company stock may be a good gamble even though the laboratory takes all the net earnings in hard times. WHEELER P. DAVEY

IRON-BORON ALLOYS and 18-8 Steels plus BORON

Their Hardening and Precipitation-Hardening

By DR. R. WASMUHT†

RECENT investigations have shown that members of certain iron alloy systems lend themselves to age hardening within a suitable alloy range. These previous investigations have dealt chiefly with either alloys of a system possessing an extended γ region (Type 3 in Wever's[1] system), such as the age hardening alloys of iron with carbon, nitrogen and copper, or else with such alloys possessing a closed γ region and but limited solubility of the alloying constituent in the *a*-iron (Type 2), such as the age hardening alloys of iron with tungsten, molybdenum, beryllium, phosphorus and titanium. In the present paper it is proposed to show that the alloys of the two types mentioned above are by no means the only ones that are capable of age hardening, and that elements whose binary alloy diagram with iron shows a contracted γ-region (Type 4) can also, with suitable treatment produce precipitation hardening in iron.

According to F. Wever and M. Mueller[2], the iron-boron system belongs to a group that exhibits a contracted γ-region (Fig. 1). The dissolving power of γ-iron for boron varies, according to temperature, from 0.10 to 0.15%. The A_3 point is raised by boron and the A_4 point lowered at the same time. Consequently, the solubility of boron is slightly greater in *a*-iron than in γ-iron. Alloys of this type were hitherto not supposed to be amenable to precipitation hardening, as the supersaturation of the *a*-iron, and consequently the possibility of producing precipitation hardening, was rendered difficult by the fact that the solubility in γ-iron was lower than in *a*-iron.

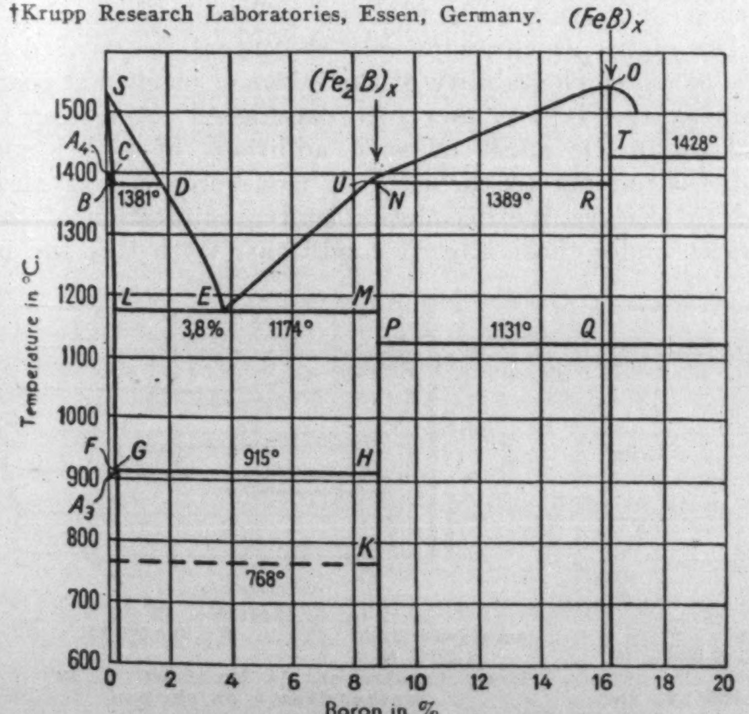

Fig. 1. The Iron-Boron Diagram

The obvious course was to artificially increase the dissolving power of γ-iron for boron by adding constituents that extend the γ region, such being chiefly manganese and nickel. By this means it must be possible under certain circumstances to extend the γ region so far that it overlaps the region of homogeneous *a*-iron and so to render such alloys amenable to precipitation hardening in a similar manner as the types of alloys mentioned above. The extension of the γ region of iron-boron alloys by means of an additional alloying constituent has already been observed. The experiments of Wever and Mueller have proved that the extension of the γ region with decreasing temperature, originally assumed by Tammann and Vogel[3] for the binary Fe-B system, is due to the presence of carbon, that is to say, of an element that also extends the γ region. In the investigations dealt with in the present paper, alloying with carbon was not resorted to, because the hardening phenomena due to the carbon would have masked the expected hardening effect of the boron.

For the purpose of these experiments, a series of iron-boron alloys were prepared,* the analysis of which is given in Table 1. In addition to binary alloys, boron steels containing also Mn, Ni-Cr, and Si were made in a high frequency furnace. Samples of these alloys were then tested for precipitation hardening by the well-known method of establishing the tempering isothermals. The samples (excepting the Cr-Ni steels, which were treated differently) were subjected to the following treatment: In the case of one series quenching was carried out below a_3 at 750° C. in water, while another series was quenched above A_3 at 1000° C. also in water. Tempering was effected at temperatures of 20°, 100°, 200°, 300°, 400°, 500°, 700° and 800° C. Previous to this treatment, all samples had been normalized by annealing for 1 hour at 930° C. and allowing them to cool

Table 1. Chemical Composition of the Boron Steels Investigated

Group	No.	C	B	Si	Mn	P	S	Cr	Ni
Not alloyed	1								
	2	0.02	0.11	0.25	0.33	0.01	0.03	......	......
	3	0.03	0.34	0.25	0.43	0.01	0.03	......	......
	4	0.02	1.34	0.36	0.41	0.01	0.04	......	......
	5	0.04	2.53	0.78	0.40	0.01	0.03	......	......
Mn additions	11	0.04	0.38	0.10	2.28	0.01	0.024	......	......
	12	0.06	0.62	0.30	2.42	0.01	0.024	......	......
	13	0.06	1.26	0.42	2.00	0.01	0.030	......	......
Si additions	21	0.06	0.11	1.97	0.51	0.01	0.024	......	......
	22	0.05	0.30	1.99	0.56	0.01	0.024	......	......
	23	0.06	1.27	1.01	0.58	0.01	0.024	......	......
	24	0.07	1.30	2.57	0.69	0.01	0.024	......	......
Ni, Cr additions	30	0.14	0.24	0.66	0.77	......	......	17.8	7.96
	31	0.16	0.37	0.71	0.27	......	......	19.8	9.7
	32	0.15	0.44	0.56	0.55	......	......	18.4	8.06
	33	0.16	0.55	0.51	0.41	......	......	17.3	7.5
	34	0.17	1.14	0.71	0.56	......	......	17.3	7.5

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†Krupp Research Laboratories, Essen, Germany.

*Ferro-boron with a content of 20% B was used. Of the boron added 70% was recovered.

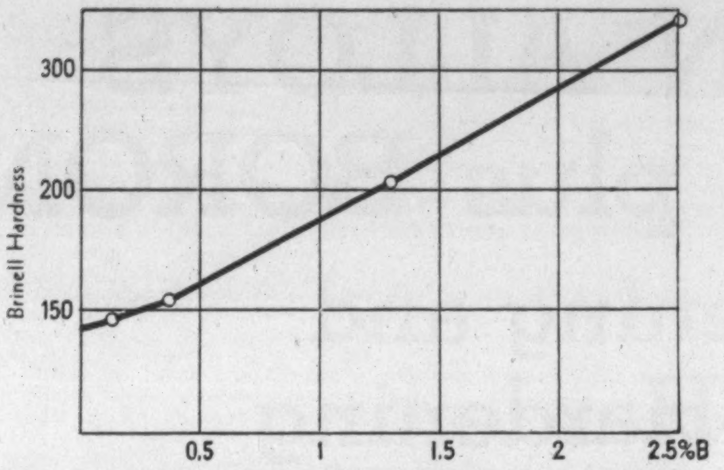


Fig. 2. Relation Between Boron Content and Hardness of Iron-Boron Alloys.

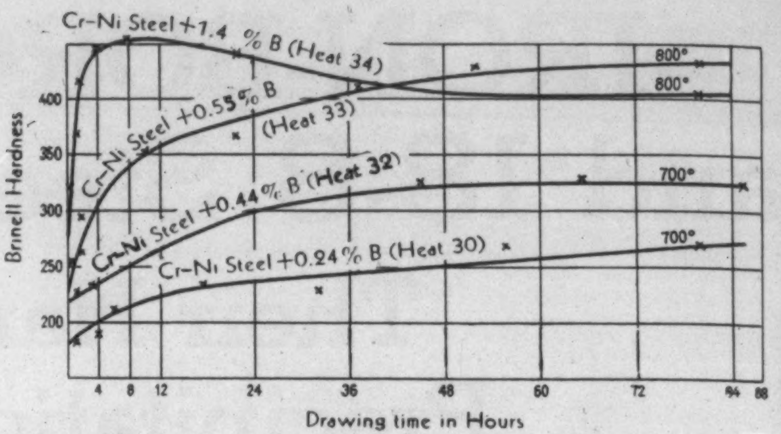


Fig. 6. Relation of Hardening in 18-8 Type of Steel to Boron Content. Heat Treatment: Quenched 1200° C. in water, drawn as indicated.

down in the furnace. In the case of casts 1 to 5, which contained no additional alloying element, but only boron contents increasing from 0.06 to 2.5%, it was found, in the first place, that the hardness of the normalised material increased with increasing boron content (Fig. 2). By the heat treatment described above, no further increase in hardness, that could be regarded as hardening in the proper sense of the term was secured, the hardness remaining practically unchanged after both the quenching and the annealing processes.

Quite different results were obtained when similar boron steels were alloyed with about 2% Mn. The analysis of these steels (samples 11 to 13) is also given in Table 1. *With these samples a pronounced quench-hardening effect was observed,* which was but slightly increased by the subsequent annealing process and could even be destroyed again completely by increasing the duration of the annealing operation or raising the annealing temperature. This process is illustrated in Figs. 3 to 5, which represent the results obtained with the samples in question when quenched from 1000° C. A significant increase of the Brinell hardness is observable similar to normal martensitic hardening. The hardness obtained ranged from 340 to 400 Brinell, the increase in hardness due to quenching amounting to about 200 Brinell. Subsequent annealing produced softening, increasingly so with increasing tempering temperature. In order to completely restore the hardened material to the soft condition, a temperature of 700° C. is generally required. Annealing at very low temperatures 100° to 150° C., however, at first caused a slight additional increase in hardness. It is known[4] that martensite also can be slightly hardened by tempering at 100° to 150° C. In the case of sample 13, which has the highest boron content, the slight additional increase in hardness at the beginning of the annealing operation could be observed up to a temperature of 400° C. (Fig. 5).

Besides being quenched above A_3, the same steels were also, as mentioned above, quenched below A_3 at 750° C. Under these conditions, no hardening effect due to quenching was obtained with any of the samples, as is also known to be the case with iron-carbon alloys. Subsequent tempering also gave no appreciable change in hardness.

The experiments described above thus show that the addition of manganese causes a widening of the γ region, or in other words, increases the dissolving power of the γ-iron for boron and so sets up conditions similar to those obtaining in the case of iron-carbon alloys, with the result that quenching produces a hardening that resembles martensitic hardening, while tempering causes a corresponding softening of the material.

Fig. 11 illustrates the structure of the quenched sample 11, containing 0.38% B. The state of strain produced by the spontaneous supersaturation of the γ-iron gives to the etching a martensitic appearance. The primary grain has remained visible in this instance, because undissolved particles of the eutectic E have been retained in the grain boundaries. It is obvious that an addition of 2% Mn was only able to increase the dissolving power of the γ-iron to a value slightly below 0.38% B, so that complete dissolution of the eutectic could not be brought about by tempering for ½ hour at 1000° C. The undissolved remnants of the eutectic therefore remained deposited along the primary grain boundaries, without participating in the hardening process.

A further element which could be expected to produce the desired widening of the γ region was nickel. It is well-known that the widening of the γ region of iron-carbon diagram can, in the case of steels containing a higher nickel or chromium-nickel content, be carried to a point at which stability of the γ-iron is reached at room temperature. It was therefore considered interesting to investigate the effect of such additions of nickel and nickel-chromium in the case of iron-boron alloys also, and to observe the dissolution and precipitation of the borides under these altered conditions. With this end in

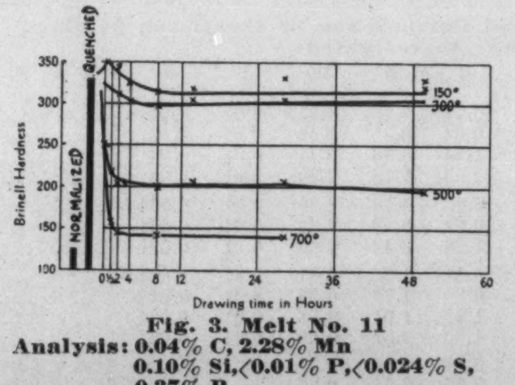


Fig. 3. Melt No. 11
Analysis: 0.04% C, 2.28% Mn 0.10% Si, <0.01% P, <0.024% S, 0.37% B
Heat Treatment: 1100° C. in water, drawn as shown.

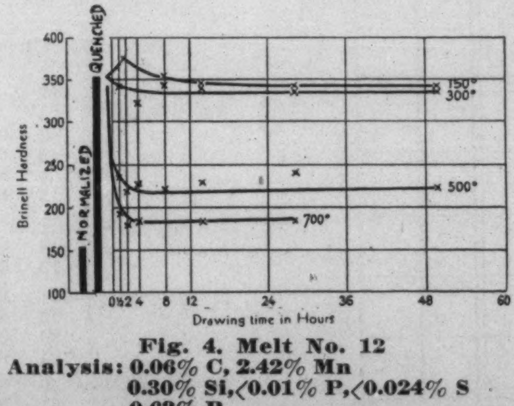


Fig. 4. Melt No. 12
Analysis: 0.06% C, 2.42% Mn 0.30% Si, <0.01% P, <0.024% S 0.62% B
Heat Treatment: ½ hr. 1100° C. in water, drawn as shown.

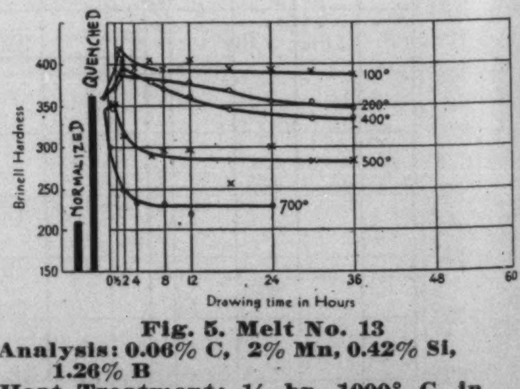
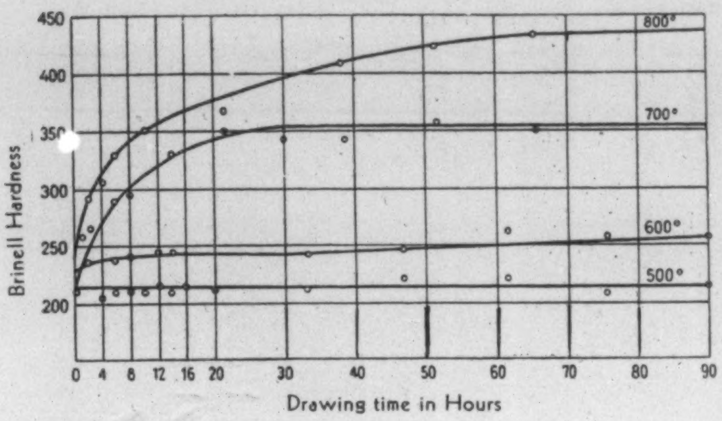


Fig. 5. Melt No. 13
Analysis: 0.06% C, 2% Mn, 0.42% Si, 1.26% B
Heat Treatment: ½ hr. 1000° C. in water, drawn as shown.

Change of Hardness of Melts 11, 12 and 13 Due to Quenching and Drawing.



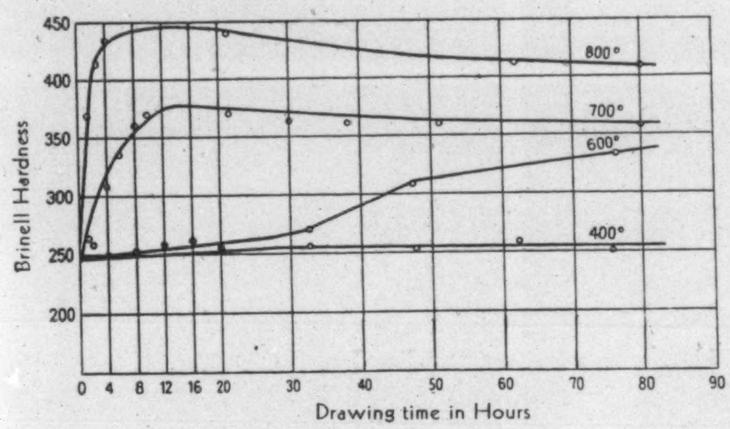
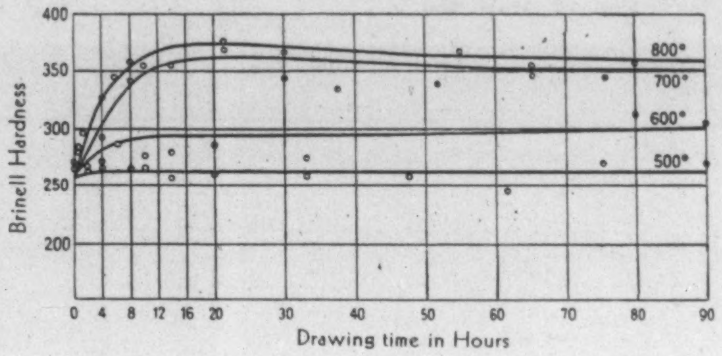


Fig. 7. (Above) Melt No. 33
Analysis: 0.16% C, 0.41% Mn, 0.51% Si, 17.3% Cr, 0.55% B
Heat Treatment: ½ hr. 1250° C. in water, drawn as shown.
Fig. 8. (Below) Melt No. 34
Analysis: 0.17% C, 0.56% Mn, 0.71% Si, 17.3% Cr, 1.14% B
Heat Treatment: Same as Fig. 7.



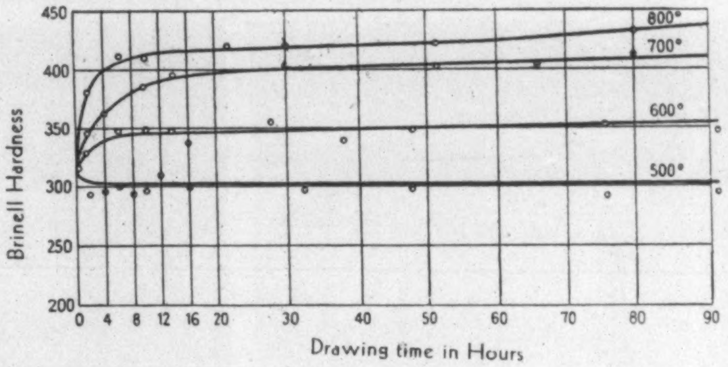


Fig. 9. (Above) Melt No. 33
Analysis: Same as Fig. 7
Heat Treatment: ½ hr. 1000° C. in water, drawn as shown.
Fig. 10 (Below) Melt No. 34
Analysis: Same as Fig. 8.
Heat Treatment: Same as Fig. 9.

+Na

Age Hardening of Melts 33 and 34

view a series of stainless nickel-chromium steels with different boron contents were made, the chemical composition of which is also given in Table 1. In this instance, quenching was effected from 1000° and 1200° C., while the subsequent tempering operation was carried out at the same temperatures as were applied in the tests previously described.

The behavior of these steels differed entirely from that of the manganese-bearing steels mentioned above. To begin with, no appreciable quench-hardening occurred with any of them. The quenched steels were comparatively soft and also all of them non-magnetic, i. e. austenitic. The only difference between the various alloys consisted in that the hardness increased slightly with increasing boron content. *Subsequent annealing, however, produced pronounced precipitation hardening in all these steels.* As will be seen from Fig. 6, the tendency to precipitation hardening perceptibly increased with increasing boron content. The higher the boron content, the more steeply does the curve rise, in other words, the earlier did hardening set in. The hardness obtained was approximately 450 Brinell, the increase due to tempering amounting to about 200 Brinell. Taking as an example the alloy containing 0.55% B, it will be seen that 80 hours were required to reach approximately the same degree of hardness as was obtained in only 8 hours in the case of the alloy containing 1.14% B. The tempering temperatures required to bring about precipitation hardening were comparatively high, no effect whatever having been observed below 600° C. The optimum tempering temperatures ranged from 700° to 800° C. The alloys having a higher boron content (from 0.5 to 1% B) should preferably be tempered at 800° C., whereas 700° C. proved to be more suitable for the alloys with a lower boron content (0.2 to 0.5% B). Some characteristic tempering isothermals for samples 33 and 34, water quenched from 1200° B, are reproduced in Figs. 7 and 8. The tempering isothermals exhibit in every particular the course characteristic of precipitation hardening.

In the samples quenched from 1000° C., subsequent tempering also produced precipitation hardening, though to a somewhat lesser extent, particularly in the case of low boron alloys. After quenching a slight increase in hardness was noticed, especially in the case of alloys with a higher boron content, the increase amounting to about 40 Brinell in the case of alloy 33 (0.55% B) and to about 60 Brinell in the case of alloy 34 (1.14% B). This is an indication that cooling did not take place quickly enough, so that part of the excess supersaturated phase could be precipitated in a highly dispersed state i. e. in a hardening manner (Figs. 9 and 10).

The precipitation hardening phenomena were also studied by the metallographical method. Figs. 12 to 17 are etched sections showing the various phases of the precipitation hardening process in the case of sample

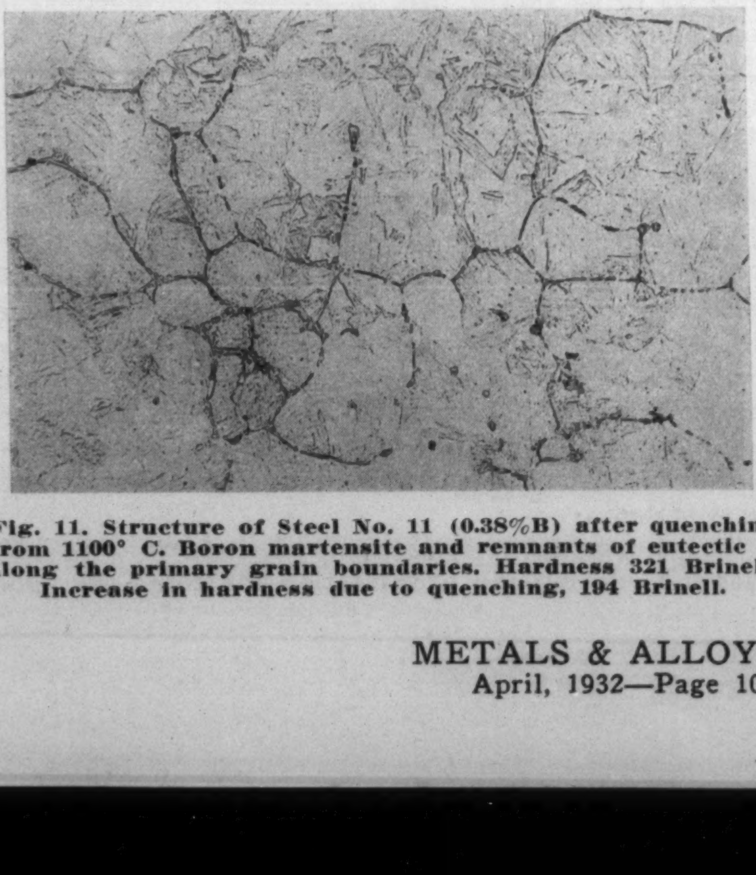


Fig. 11. Structure of Steel No. 11 (0.38%B) after quenching from 1100° C. Boron martensite and remnants of eutectic E along the primary grain boundaries. Hardness 321 Brinell. Increase in hardness due to quenching, 194 Brinell.

Fig. 12. Mag. 200 ×

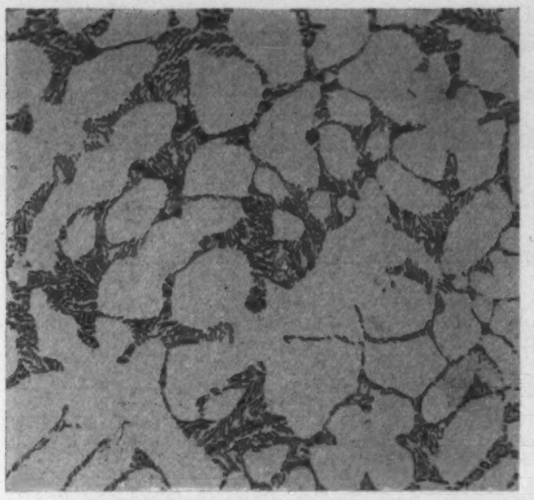


Fig. 14. Mag. 200 ×

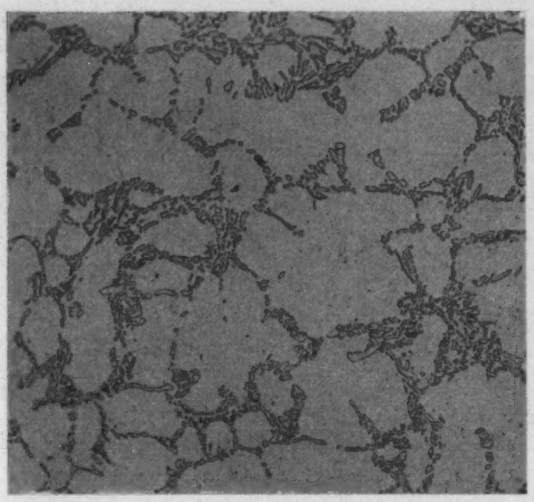
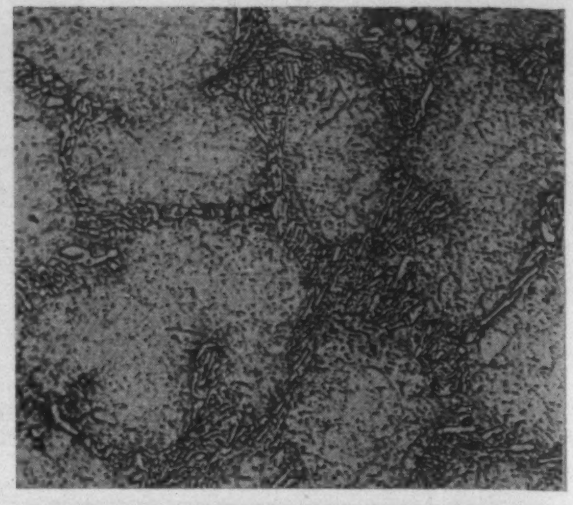


Fig. 16. Mag. 500 ×



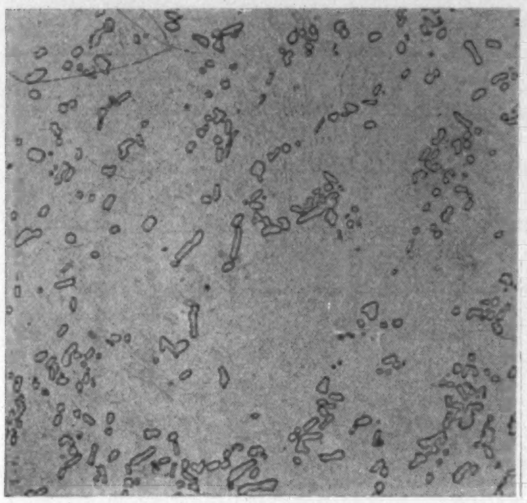


Fig. 13. Mag. 200 ×
As Supplied

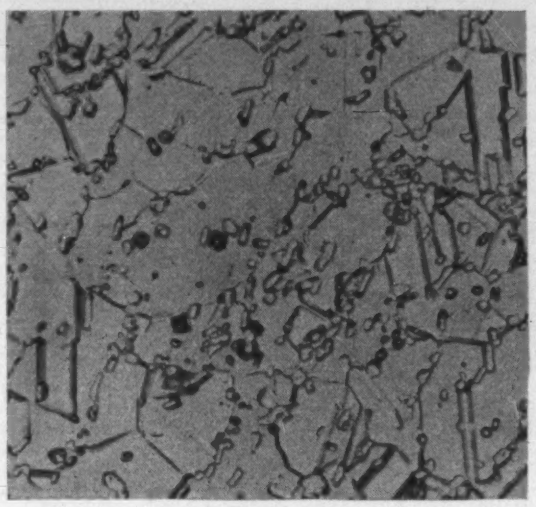


Fig. 15. Mag. 200 ×
Quenched from 1150° C.

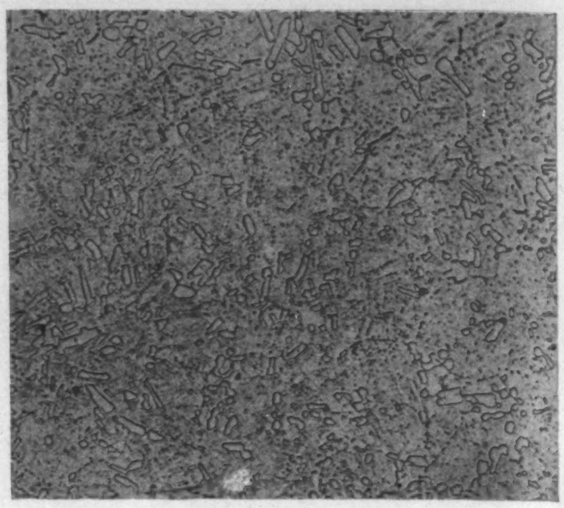


Fig. 17. Mag. 200 ×
Age-hardened at 1150° C./800° C.

STEEL No. 33(0.55%B) { **Figs. 12, 14, 16—CAST STATE**
{ **Figs. 13, 15, 17—FORGED STATE**

33 (0.55% B). In the cast state the structure shows precipitations of an eutectic corresponding to Ledeburite (Fig. 11) which was coagulated to a certain degree by the forging process (Fig. 13). A certain part of this eutectic could be dissolved by quenching (Fig. 14), which operation, especially in the forged sample, further clearly revealed its austenitic character. Subsequent tempering resulted in precipitation of the excess part in a finely dispersed i. e. hardness-producing state, as indicated by the minute dots appearing between the undissolved coarse precipitations. These dots were not visible at the time when the increase in hardness began; only when the age hardening had been carried just beyond the maximum did they become visible in the manner shown in Figs. 16 and 17. The rate of growth of the boride particles causing hardening by their precipitation appears to be slower than in the case of precipitation hardening produced by the addition of titanium, where the precipitations become visible shortly after the increase in hardness sets in. Even in the age hardened state the samples remained non-magnetic, i. e. austenitic. It was not until the conclusion of the annealing process, shortly before the precipitations became visible, that a slight magnetisation of the sample could be detected.

Finally, some boron steels containing silicon (Table 1) were also made, in order to investigate the influence of silicon on the hardening of iron-boron alloys. Their behavior on treatment was found to resemble that of the binary boron steels discussed above in the first place, that is to say that no appreciable precipitation-hardening or quench-hardening took place within the analysis limits investigated.

The foregoing investigations have shown that both quench-hardening and precipitation-hardening are liable to occur with iron-boron alloys. The two kinds of hardening apparently differ only in so far as the degree of supersaturation is concerned. We have seen that the binary iron-boron alloys did not show signs of hardening, as, owing to their contracted γ region, no marked supersaturation could take place. By adding a few percent of manganese it was possible to widen the γ region so far that it exhibited conditions similar to those obtaining in the case of iron-carbon alloys having a widened γ region. In any case, the widened γ region notwithstanding, the γ phase formed remained instable at low temperature. Decomposition had to take place with formation of a-iron and precipitation of the excess boride owing to the lower dissolving power of the γ-iron. These manganese bearing alloys thus show an abrupt and discontinuous decrease in dissolving power during the γ-a transition. This causes a very strong supersaturation of the a-iron in quenching, which results in quench-hardening of the martensitic type. In the third series of samples, consisting of boron steels containing an appreciable amount of nickel and chromium-nickel, the widening of the γ region could be carried so far that the γ phase became stable even at room temperature, in other words, that an open γ region was formed. As already mentioned, the steels in question were all non-magnetic. Consequently no abrupt and considerable changes in dissolving power, such as occur during the transition from the γ to the a phase, have to be overcome in cooling. The supersaturation with boride on quenching keeps within comparatively narrow limits. The pressure of the phase that tends to precipitate is not high enough to cause precipitation at once. The latter must, on the contrary, be started by the annealing process. Accordingly, hardening only takes place during the annealing process.

INFLUENCE OF THE RATE OF COOLING DURING THE QUENCHING PROCESS ON THE HARDENING PHENOMENA OF IRON-BORON ALLOYS

An interesting insight into the relationship between quench-hardening and precipitation-hardening may be gained by considering the results of experiments undertaken with a view to determine the influence exerted on the hardening phenomena of boren steels by the rate at which they are cooled from elevated temperatures. A series of the alloys discussed above was cooled at different rates from the quenching temperature and subsequently subjected to the known annealing process. The rate of cooling was varied by employing various quenching media, *i. e.* water, oil, lead and air at rates of 8000°, 1000°, 500°, 300°, 200°, 100°, 50°, 25° and 10° per hour. The results of these experiments are recorded in Fig. 18. The alloy selected as examples is 12, representing the quench-hardening type of boron steels and 33, representing the precipitation-hardening type of boron steels.

In the case of melt 12 it was found that the highest hardness was secured by the most drastic cooling medium. By air cooling, considerable hardness was still obtained in some instances, but the hardness decreased in a marked manner with the decrease of the rate of cooling. Complete absence of hardening effect was only noted when the rate of cooling had been reduced to 10° per hour, the sample retaining approximately the hardness which it had possessed in the normalised state. Thus, the slower-acting the cooling agent, the more time is allowed for the particles precipitated during the dissociation to ball up into coarser coagulated particles which no longer produce any hardening effect.

Conversely, in the case of the age-hardening chro-

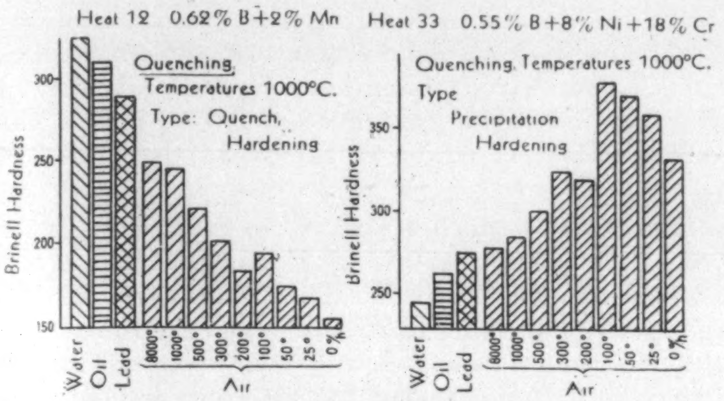


Fig. 18. Influence of Rate of Cooling from 1000° C. on the Hardening of Various Boron Steels.

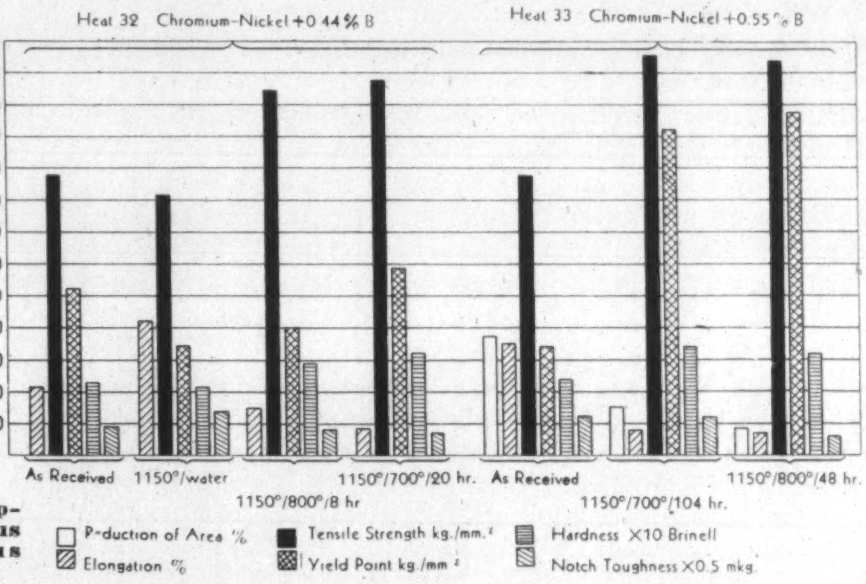


Fig. 19. Physical Properties of 18-8 Steel plus Boron After Various Heat Treatments.

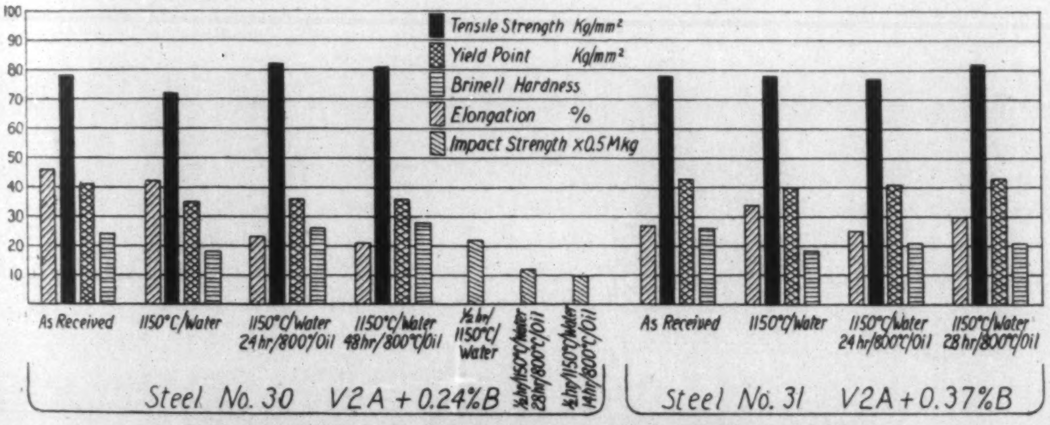


Fig. 20. Physical Properties of 18-8 Steels with less than 0.4% B After Various Heat Treatments.

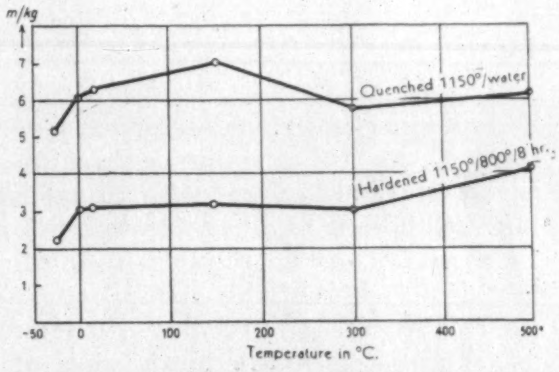


Fig. 22. Impact Strength-Temperature Curves of 18-8 Boron Steel No. 32 (0.44% B).

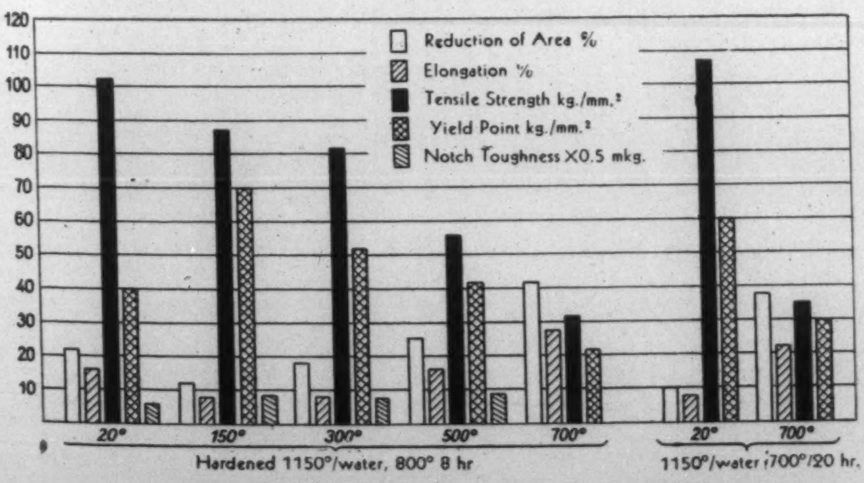


Fig. 21. High-Temperature Tensile Tests with Age Hardened 18-8 Boron Steel (Steel No. 32, 0.44% B); Time of Test 20 mins.

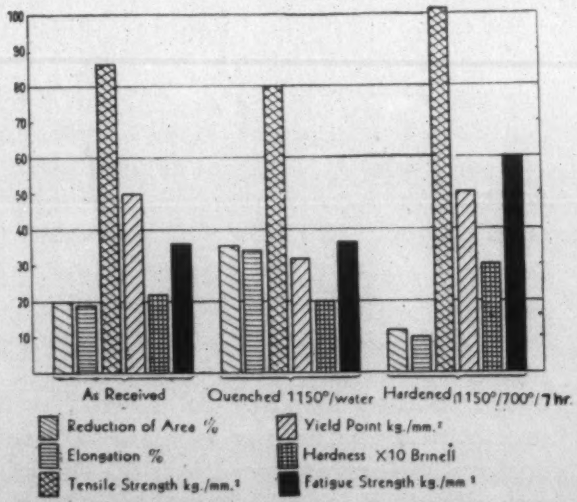


Fig. 23. Fatigue Strength Under Alternating Bending Stress of 18-8 Boron Steel (0.44% B) After Various Heat Treatments.

mium-nickel-boron steel of melt 33, the samples subjected to the most drastic quenching showed the least hardness. Even those that had been cooled rapidly in air were comparatively soft. With decreasing rate of cooling, however, a distinctly perceptible increase in hardness set in. The sample cooled at the rate of 100° per hour showed pronounced signs of hardening. Further retardation of the cooling process again produced gradual softening which is probably due to the same causes as the softening that occurred with the slowly cooled samples of the previously discussed melt 12. On being subsequently tempered, the samples that had been subjected to rapid cooling (from water cooling down to air cooling at the rate of 200° per hour) proved capable of precipitation hardening. The most drastically quenched samples showed the highest degree of hardening, whereas the slowly cooled samples had lost the capacity for precipitation hardening. It would appear, then, that the more drastic the quenching medium employed, the greater is the amount of excess boron that is held in solution and on tempering precipitates in the dispersed state, i. e. with a hardening effect observed with a certain rate of cooling (100° per hour in the case under consideration) is probably brought about in the following manner: On the one hand, the rate of cooling is not high enough to hold the excess boride in solution, so that it was able to separate out in a highly dispersed state, i. e., with a hardening action. On the other hand, the rate of cooling is not slow enough to perhaps allow the particles that have separated out in the highly dispersed state to coagulate and so again destroy the hardness previously caused by them. In fact, such subsequent softening only occurred at still lower rates of cooling. In the latter cases, annealing also merely resulted in further softening. In this melt we therefore have *an alloy that lends itself both to precipitation-hardening and to quench-hardening,* which proves that the two processes are in principle identical.

SOME PHYSICAL PROPERTIES OF AGE-HARDENING CHROMIUM-NICKEL-BORON STEELS

As appears from the foregoing experiments, particular interest attaches to stainless chromium-nickel steels that have been rendered amenable to age hardening by alloying with boron. In what follows, it is proposed to give a short survey of the physical properties that can be attained in such steels by precipitation hardening. The experiments were carried out with samples of melts 30 to 33, referred to above. In the case of steels with a low boron content (up to 0.37%, melts 30 and 31) no marked increase in tensile strength and yield point due to the aging treatment itself could be ascertained (Fig. 20). For all that, the yield point even of steels containing less than 0.4% B is considerably higher, in all the samples investigated, than that of V2A steel containing no boron, amounting as it does to 49,770-56,880 lbs./in.^{2}, with a tensile strength of 99,540-113,760 lbs./in.^{2} A marked rise in tensile strength and yield point due to the aging treatment occurs only in V2A steels having a boron content of over 0.4%. Fig. 19 shows the change of physical properties during the different phases of the aging treatment in the case of melts 32 and 33, containing 0.44 and 0.55% B, respectively. Both steels in the forged state already showed comparatively high tensile strength, about 127,980 lbs./in.^{2} In steel 32, a certain amount of precipitation hardening appears to have already occurred during the forging operation, for the yield point was also rather high, about 75,370 lbs./in.^{2} Quenching from 1150° obviously caused renewed dissolution of the precipitated particles; the yield point dropped to 49,770 lbs./in.^{2} with a slight decrease in hardness, from 228 to 212 Brinell. Elongation and impact strength at the same time rose. After tempering the quenched samples for a short time at 800° C., or for a longer time at 700° C., a considerable increase in tensile strength and yield point was noted. In the case under consideration, the tensile strength of steel 32 rose to 167,800 lbs./in.^{2} and the yield point to 83,900 lbs./in.^{2} The increase in tensile strength was accompanied by an increase in hardness amounting to about 100 Brinell. Elongation and impact strength diminished as a result of the annealing. The same phenomena, but more markedly, occurred in the case of steel 33, containing a higher percentage of boron (0.55%). Here the higher boron content caused a conspicuous increase in yield point; while the tensile strength increased to 179,170 lbs./ in.^{2}, the yield point actually rose to above 142,200 lbs./in.^{2} The increase in hardness, as well as the drop in elongation and impact strength, could be clearly observed. The impact strength of the age hardened boron-bearing V2A steels containing more than 0.4% B varied between 18 and 29 ft.-lbs., whereas V2A steels containing less than 0.4% B showed impact strengths varying from 58 to 36 ft.-lbs. in the aged state and from 87 to 72 ft.-lbs. in the quenched state.

Rupture tests in the hot state were carried out with steel 32 (V2A containing 0.44% B). Fig. 21 shows the results obtained with age hardened specimens having undergone different aging treatments. The drop in tensile strength was comparatively slow. The specimen quenched from 1150° C. and annealed for 20 hours at 700° C. still showed at the latter temperature a tensile strength of 51,190 lbs./in.^{2} with a yield point of 42,660 lbs./in.^{2}

As already mentioned, the aging treatment resulted in a lowering of the impact strength, such as has so far been observed with all age hardening alloys. Fig. 22 gives the impact strength-temperature curves for steel 32. At elevated temperature a marked drop in impact strength was noted in the age hardened specimen, as compared with the quenched specimen.

In conclusion, mention will be made of some tests made for determining the fatigue limit under alternating bending stress of age hardened boron-bearing V2A steels. The results of these tests, which were again carried out with steel 32 (0.44% B) are given in Fig. 23. Schenck's bending machine was used for the purpose. Tensile test pieces were taken from material immediately adjoining the bending test pieces and were subjected to the same treatment; the results obtained with these are included in Fig. 23. It was found that the fatigue limit under alternating stress was markedly improved by the aging treatment. The age hardened test pieces gave a value of 85,320 lbs./in.^{2} with a tensile strength of about 142,200 lbs./ in.^{2}, a result that must be regarded as excellent.

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